

B. ZAKHAROV

HEAT
TREATMENT
OF
METALS

B. Z A K H A R O V

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OF
METALS

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CONVENTIONAL SYMBOLS

Value	Letter symbol	Unit of measurement
Tensile strength (ultimate strength)	σ_t	kg/mm ²
Yield point (yield strength), physical or conventional, defined as the stress that will cause the permanent set in a specimen, equal to 0.2 per cent of its gauge length. The yield strength is characteristic of the metal elastic properties .	σ_y	kg/mm ²
Per cent elongation, when the ratio of the specimen gauge length l to its diameter d is equal to l:d = 5	σ_s	%
Per cent reduction of area (contraction)	ψ	%
Impact strength, determined by testing the specimens of Menager type	a_t	kgm/cm ²
Brinell hardness number	H _B	kg/mm ²
Rockwell "C" hardness number (C scale)	R _C	—
Temperature	t	°C
Absolute temperature	T	°C

For the above listed properties and values no units of measurement are indicated in the tables and figures.

PART ONE

TRANSFORMATIONS IN STEEL DURING ITS HEAT-TREATMENT

Chapter I

GENERAL DATA ON STEELS

1. CLASSIFICATION OF STEELS

Steels are alloys, the essential ingredients of which are: a) iron, and b) carbon in quantity not exceeding 2 per cent.

High-alloy complex products, composed of iron, chromium, nickel and other elements, in which the maximum content of carbon is limited to very low values (0.10 to 0.12 per cent), are also called steels. In these steels, carbon may be considered essentially as an undesirable impurity, but the number of such steels is very small. For the overwhelming majority of steels carbon is the second essential ingredient after iron.

Steels are also classified in two groups, according to the uses for which their properties fit them, i. e., structural steel, and tool steel. Structural steels with special physical and physico-chemical properties may be subdivided into separate groups embracing corrosion-resistant, acid-resisting, non-scaling, heat-resisting, electrical, magnetic, low-magnetic (non-magnetic), and wear-resisting steels.

Tool steels differ from structural ones by their higher carbon content. The carbon content in structural steels usually does not exceed 0.7 per cent; the majority of tool steels contains more than 0.7 per cent carbon, but in some highly alloyed tool steels the proportion of carbon may be less than this figure.

All steels are classified according to their *chemical composition* as *carbon steels* or *alloy steels*. Carbon steels are those containing only two elements: iron and carbon. Such elements as silicon, manganese, phosphorus and sulphur, contained in each steel, including those of the carbon type, are considered impurities, not ingredients. Steels containing up to 0.2 per cent carbon are listed by the U.S.S.R. State Standard in the group of low-carbon steels; those with a carbon range from 0.25 to 0.6 per cent in the group of medium-carbon

steels, and steels with a carbon content above 0.6 per cent in the group of high-carbon steels.

Alloy steels are those which, in addition to the essential ingredients, i. e., iron and carbon, contain special alloying elements—chromium, nickel, tungsten, molybdenum, vanadium, aluminium, titanium, etc. In some cases silicon and manganese are considered the alloying elements, whereas chromium, nickel and other typical alloying elements may be considered impurities.

The silicon content, as an impurity, in plain carbon and in most low-alloyed steels should not exceed 0.4 per cent. In some high-alloy steels the maximum silicon content is 0.6 per cent. In steels alloyed with silicon the proportion of this element, in general, exceeds 0.8 per cent. As is clear from the above data, it is impossible for all grades of steel to state precisely up to what proportion silicon is to be considered as an ordinary impurity, and above what figure it becomes an alloying element. But conditionally one may accept the following figures as valid for the majority of steels; silicon up to 0.8 per cent is considered an impurity, and over 0.8 per cent—an alloying element.

Similarly, if the whole totality of steel is considered, it cannot be precisely stated up to what proportion manganese is an impurity, and above what figure it becomes an alloying element. One can only accept, in general, that manganese up to 1 per cent is an impurity, and over that figure—an alloying element.

Almost all types of carbon steel contain chromium and nickel. These elements become ingredients of steel due to the usage of scrap alloy steel in charges for melting furnaces. State Standards for plain carbon steels specify the permissible contents of chromium and nickel as not exceeding 0.30 per cent of each. In chromium alloyed steels the minimum content of this element is as much as 0.40 per cent. The minimum proportion of nickel in steels alloyed with this element is 0.50 per cent.

For any grade of steel it can very easily be determined whether a given element is an impurity or an alloying element if a standard steel grade designation is considered. But this is possible only in cases when the steel grade designation is composed of the letter symbols for elements. Impurities may also be distinguished from alloying elements by the fact that in specifications of the chemical composition of steel only the upper limits of content of the former are indicated (e. g., ≤ 0.30 per cent), whereas for the latter the lower limits are indicated as well (e. g., 0.50 to 0.90 per cent). This rule, however, does not apply to silicon and manganese, for which elements, if they are indeed considered as impurities, both the upper and lower limits are indicated in specifications. For instance, the silicon content of most carbon steels is accepted as ranging from

0.17 to 0.37 per cent. This is caused by intentional adding of silicon and manganese (in the case of rimmed steel only manganese is added) to the steel for its deoxidation; therefore presence of these elements in steel testifies to its having been deoxidised.

By combining these two classifications (namely, according to purpose and to chemical composition) in one common classification, the following five groups of steel have been obtained:

- 1) carbon structural steels;
- 2) carbon tool steels;
- 3) alloyed structural steels;
- 4) alloyed tool steels;
- 5) highly alloyed structural steels with special physical and physico-chemical properties.

Apart from the classification of steels according to their purposes and chemical composition, they are also classified according to their quality characteristics as:

- 1) ordinary quality steels;
- 2) higher quality steels;
- 3) quality steels;
- 4) high-quality steels.

The distribution of the above five groups of steel according to quality is given in Table 1.

Table 1

Classification of Steels According to Quality
and Methods of Manufacture

Ordinary and higher quality steels	Quality steels	High-quality steels
	Carbon structural steels	Carbon tool steels
	Alloyed structural steels	Alloyed tool steels
		Highly alloyed structural steels with special properties
Open-hearth steels Bessemer steels	Open-hearth steels Electric furnace steels	Electric furnace steels Open-hearth steels

According to State Standards alloyed steels are classified as: *low-alloy* steels, if the total (summary) amount of alloying elements contained does not exceed 2.5 per cent; *medium-alloy* steels, if this amount is in the range of 2.5 to 10 per cent; and *high-alloy* steels, if it exceeds 10 per cent.

The qualitative class of steel is determined by the degree of its purity with regard to harmful impurities and foreign non-metallic matters (inclusions), the uniformity of its chemical composition, and, as a consequence, the higher values of its mechanical and physico-chemical properties. The lower the permissible contents of sulphur and phosphorus in steel, the higher is the class of its quality (see Table 2).

Table 2

Permissible Contents of Sulphur and Phosphorus in Steel

Group of steel	Permissible content in per cent	
	Sulphur	Phosphorus
Carbon structural steels of ordinary quality		
open-hearth steels	0.055	0.050
Bessemer steels	0.060	0.080
Higher quality and quality steels	0.040	0.040
Carbon tool steels		
quality steels	0.030	0.035
high-quality steels	0.020	0.030
Alloy structural steels		
quality steels	0.040	0.040
high-quality steels	0.030	0.035
Alloy tool steels	0.030	0.030
High-alloy structural steels with special properties	0.030	0.035

In analysing figures given in Table 2 a question may often arise: is it really possible that such a negligible difference (0.01 per cent) in sulphur and phosphorus contents of quality and high-quality steels can bring about an appreciable change in properties of these steels? Such a change, of course, is caused not only by the difference in sulphur and phosphorus contents. The sulphur and phosphorus proportions are factors characteristic of purity of steel, factors that can easily be determined by chemical analysis. As a rule, it may be accepted that if the contents of sulphur and phosphorus in steel are low it will also be cleaner with regard to other

harmful impurities (oxygen, nitrogen, etc.), which are difficult to determine analytically, and especially with respect to solid non-metallic inclusions.

The specified purity of steel with regard to harmful impurities and non-metallic inclusions can be achieved by proper selection of charging materials and of method of manufacture. The highest purity is obtained by melting steel in electric furnaces, the lowest purity—by making it in Bessemer converters. On the other hand, electric furnace steel is the most expensive, whereas Bessemer steel is the cheapest. Selection of manufacturing methods for steels of various quality classes is made on the basis of these considerations. Structural steels of *ordinary quality* are made in open-hearth furnaces or in Bessemer converters. Higher quality steels are almost exclusively made in open-hearth furnaces. *Quality* steels are, in general, manufactured in open-hearth furnaces and, sometimes, in electric furnaces. *High-quality* steels, on the contrary, are generally made in electric furnaces; only in special cases can they be melted by the open-hearth process.

The letter A is appended to the specification number designating the grade of a high-quality steel in those cases when the same grade may be classed either as a quality or as a high-grade steel.

2. CARBON STEELS AND THEIR CHARACTERISTICS

Structural carbon steels. Structural carbon steels are divided into three groups according to quality: *ordinary quality*, *higher quality* and *quality* steels.

The chemical compositions and properties of *structural carbon steels* of ordinary and higher quality types are indicated in the State Standards. According to these Standards all structural carbon steels are divided into three groups:

a) In the first group are steels with guaranteed mechanical properties; their chemical composition is optional (not obligatory), except for sulphur and phosphorus; the main grades of this group are designated as follows: BCr. 3, BCr. 4, BCr. 5, BCr. 6 (Bessemer steels), and Cr. 1, Cr. 2, Cr. 3, Cr. 4, Cr. 5, Cr. 6, Cr. 7 (open-hearth steels).

b) Steels of the second group have guaranteed chemical composition; the main grades of this group are designated as follows: B09, B10, B23, B33 (Bessemer steels), and M09 кп, M12 кп, M18 кп, M18, M21, M26, M31, M44, M56 (open-hearth steels); the numbers indicate the average carbon content in "points", or hundredths of 1 per cent, and the letter symbols "кп" designate rimmed steels (specification numbers without "кп" indicate killed steels).

c) Steels of the third group (higher quality steels) have guaranteed compositions and mechanical properties and are designated as follows: M09, M12, M16, M18a, M21a, M26a, M31a, M44a, M56a.

The higher the number designating the grade of carbon steel of ordinary quality belonging to the first group, the higher is its carbon content; thus, the carbon content in steel of grade Cr. 1 is not above 0.12 per cent, in steel of grade Cr. 4 up to 0.25 per cent. The figures 1, 2, 3, etc., in standard designations of these grades indicate only the conventional number of steel. For example, the figure 3 in designation of grade Cr. 3 does not at all signify, as is sometimes wrongly believed, that this steel contains 0.3 per cent carbon: the actual carbon content in steel of this grade is not over 0.22 per cent.

The greater the proportion of carbon in steel, the higher are the values of its tensile strength, yield point and hardness, and the lower those of elongation and resilience (impact strength). In other words, the greater the carbon content in steel, the higher, all other things being equal, are its strength, elasticity (determined by yield point) and hardness, and the lower its ductility (characterised by elongation) and impact strength.

The technological and mechanical properties of these steels depend largely upon their carbon content. Cold workability and weldability rapidly deteriorate with an increase in carbon. Steels of grades M09кп-M18кп usually are used to manufacture parts by cold working methods, whereas the steel of grade M18 is used for welded structures. An increase in carbon content also rapidly impairs the flame cuttability of the steel. Only the grades Cr. 1-Cr. 4 and M09 кп-M31 can easily be flame-cut. On the other hand, the ability of steel to harden, i. e., its liability to increase in hardness through hardening treatment, improves greatly with an increase in carbon content.

The uses of steels of the first and second groups are determined basically by the following considerations. Steels of the second group are generally used for welded structures because of their guaranteed chemical composition, since the latter (or, more precisely, the carbon content) largely determines the weldability. Therefore, the main part of welded structures is made of steel M18, but not of grade Cr. 3. Parts undergoing the strengthening heat-treatment (normalising, and hardening with subsequent tempering) are also, in general, manufactured of steels of the second group, since to have certain mechanical properties steel must have a certain carbon content. Parts which must possess certain mechanical properties and which are manufactured of rolled steel without subsequent heat-treatment, i. e., parts with mechanical properties determined by the mechanical properties of rolled stock, are usually made of steel of the first group.

Structural carbon steels of ordinary quality are generally used for making hot-rolled, so-called ordinary stock: plates, sheets, rods, channels, beams, angles, tubes, wire bars, and light-weight general-purpose forgings. Steels of ordinary quality are not used for manufacture of cold-drawn (sized) steel rods and wire.

The chemical composition and mechanical properties of carbon quality structural steels are specified by the U.S.S.R. State Standards. Quality structural carbon steels are divided into two groups according to their manganese contents: the first group with normal manganese content (not exceeding 0.80 per cent), and the second one higher in manganese (minimum manganese content 0.70 per cent).

Steels of the first group are designated by two-digit numbers indicating the average carbon content in "points" or hundredths of 1 per cent as: 05 кп, 05, 08 кп, 08, 10 кп, 10, 15 кп, 15, 20 кп, 20, 25 . . . 85 (23 grades in all). Grades 05 кп, 08 кп, 10 кп, 15 кп and 20 кп are known as rimmed steel (deoxidised only by manganese), and other grades as killed steel (deoxidised by silicon and manganese). Rimmed steel is characterised by its very high ductility and is especially suitable for deep drawing operations.

Steels of the second group (17 grades in all) are higher in manganese content and are designated by two-digit numbers indicating the average carbon content in hundredths of 1 per cent, and by the letter symbol "Г" indicating manganese: 20Г, 65Г, etc. In the designation of eight of the grades of the second group, the figure 2 is annexed to the letter symbol (10 Г2, 30Г2, etc.); the average manganese content in steels of these grades is 1.6 per cent.

Quality structural carbon steels are characterised according to quality specifications as follows:

- 1) the mechanical properties of these steels are somewhat higher than those of ordinary steels (Table 3);

- 2) quality steels are specified either by chemical composition or by mechanical properties.

Quality structural carbon steels are used for plates (both hot-rolled and cold-rolled), rods and tubes (hot-drawn and cold-drawn), wire, and special purpose forgings.

Structural carbon steels also include the so-called *free-cutting steels* that are used for small-size machine parts (in particular for various fastenings, studs and screws), especially for those produced by automatic machining. To be suitable for high-speed machining operations, these steels should possess high *machinability*, i. e., free machining properties. The poor machinability of plain low carbon steels (not free-cutting steels) is caused by their high ductility and toughness. When ordinary steels are machined, the chip comes away in the form of a long continuous helix, which causes friction and consequent intensive heating and wearing of

Table 3

**Comparison Between the Specified Mechanical Properties
of Ordinary Quality and Quality Structural Carbon Steels**
(steels of approximately the same carbon content are compared)

Grades to be compared		Tensile strength		Yield point, not less than		Elongation, not less than	
Ordinary	Quality	Ordinary	Quality	Ordinary	Quality	Ordinary	Quality
Cr. 1	10	32-40	36-45	—	22	33	32
Cr. 3	20	40-50	44-54	25	26	25	26
Cr. 5	30	50-62	50-62	29	30	19	22
Cr. 6	45	60-72	64-76	32	36	14	17
Cr. 7	55	≥70	71-83	35	40	10	13

the cutting tool. Therefore when ordinary steels are machined, reduced cutting speed and frequent change of tools are necessary, and these decrease the productivity of any machine tool, especially an automatic one.

To increase the machinability of low-carbon steels it is necessary to reduce their ductility. In this case there will not be a long continuous helical chip but a breaking off (loose) chip, and the friction between this and the cutting tool will be greatly reduced. The addition of considerable amounts of sulphur and phosphorus (up to 0.30 and 0.15 per cent respectively) in free-cutting steels reduces the ductility and thus improves the machinability of low-carbon steels. The U.S.S.R. State Standard specifies four grades of free-cutting steel: A12, A20, A30 and A40Г. In spite of the increased content of such harmful elements as sulphur and phosphorus, the mechanical properties of free-cutting steels are not altogether bad: the tensile strength and hardness of these steels are somewhat higher, elongation slightly less, and only the reduction of area considerably less (by $1\frac{1}{2}$ to 2 times) than those of ordinary and even quality steels.

This is explained as follows. Sulphur, if it is contained in the form of FeS distributed in the metal structure as long inclusions at the grain boundaries, has a very injurious effect upon the properties of the steel. But when sulphur is contained in the steel structure in the form of double sulphide MnS-FeS as small globules the harmful effect is considerably reduced. In free-cutting steels the manganese content is intentionally increased in order to ensure the forming of these globules of double sulphide (MnS-FeS).

Carbon Tool Steels. The chemical composition of *carbon tool* steels is specified by the U.S.S.R. State Standards. There are

available eight grades of quality tool steel: Y7, Y8, Y8Г, Y9, Y10, Y11, Y12, Y13 and eight grades of high-quality tool steel: Y7A, Y8A, Y8ГA, Y9A, Y10A, Y11A, Y12A and Y13A. The letter symbol Y designates carbon steel; the figures annexed to letter symbol Y indicate the carbon content in tenths of 1 per cent. The letter symbol Г, designating manganese, indicates higher manganese content; the symbol A at the end of a grade designation number indicates, as usual, that the steel belongs to the high-quality class.

The uses of tool carbon steels are determined by their carbon contents. The higher the carbon content, the greater the hardness of steel in as-hardened condition and the lower its impact strength. Therefore grades Y7 and Y7A are used for fitter's tools, which are subjected to shocks during use (point-tools, chisels, marking irons, hammers, sledge hammers, etc.). Tools, which are subjected to shocks in service and at the same time require higher hardness, such as matrices, dies, punches, shear blades for metal cutting, pneumatic chisels, etc., are made from steel of grades Y8, Y8A, or less often of grades Y8Г, Y8ГA, Y9, Y9A.

Grades Y10, Y10A, Y11, Y11A, Y12 and Y12A are used primarily for such cutting tools as drills, milling cutters, breaches, reamers, taps, files, etc., and for various gauges. Grades Y13 and Y13A are used for tools, which are not subjected to shocks and should possess very great hardness, such as chisels for cutting files, scrapers, drawing dies, boring tools, etc.

3. ALLOY STEELS AND THEIR CHARACTERISTICS

For most of grades of alloyed steels the following standard system of designations is adopted in the U.S.S.R.

Grades of *alloy structural* steels are designated by two-digit numbers indicating the carbon content in *hundredths* of 1 per cent, and by one or more capital letters indicating the alloying elements contained in the given steel. These letter symbols have the following meaning:

X — chromium	B — tungsten	T — titanium
H — nickel	Φ — vanadium	Б — niobium
Г — manganese	M — molybdenum	Ю — aluminium
С — silicon	K — cobalt	

If the content of alloying element in steel is equal to or less than 1 per cent, the corresponding letter symbol is usually not followed by a number, but if the alloying element is above 1 per cent, the letter symbol is followed by a figure indicating the *approximate* content of this element in per cent. However, it is necessary to be rather

cautious as to the latter: in some cases the real content of alloying element differs from that indicated in the grade designation number. For instance, in steel of grade 40HM the nickel content is 1.50-2.00 per cent, and in grade 20X2H4 it ranges from 3.25 to 3.75 per cent.

Grades of *alloy tool* steel are designated by one-digit numbers (i. e., by one figure) indicating the carbon content in *tenths* of per cent, and by one or more letter symbols indicating the alloying elements contained. The meaning of the letter symbols is the same as for the alloy structural steels. If the carbon content in alloy tool steel is 1 per cent or more, the numbers are not put before the letter symbols. For example, the carbon content in steel of grade XBF ranges from 0.90 to 1.05 per cent.

The same index system is used for designation of *high-alloy structural* steels with special physico-chemical properties; but in the majority of cases the numerals indicating carbon content are not put before the letter designation. The figures (one-digit) are used only in the designations of those grades of steel that differ from each other in carbon content, but not in alloying elements. For instance, four grades of stainless steel alloyed with equal proportions of chromium (12 to 14 per cent) are designated as follows: 1X13 (carbon content under 0.15 per cent), 2X13 (carbon content from 0.16 to 0.24 per cent), 3X13 (carbon from 0.25 to 0.34 per cent), and 4X13 (carbon from 0.35 to 0.45 per cent).

Apart from these simple and rational specifications, there are other designations for some groups of steel. Thus, two standard grades of *high-speed* steel are designated by the letter symbol "P" (derived from the English word "rapid"), with a subsequent numeral (9 or 18) indicating content of tungsten, which is the predominant alloying element in this kind of steel: e. g., P9 and P18.

Grades of steel used for *ball bearings* are designated by letter symbols ШХ (derived from the Russian words "ball bearing chromium steel"), followed by figures indicating the approximate chromium content in tenths of per cent, e. g., ШХ6 (chromium content from 0.4 to 0.7 per cent), ШХ9 (chromium from 0.9 to 1.2 per cent), ШХ15 (chromium from 1.3 to 1.65 per cent).

Grades of steel for *electrical purposes* (transformer and dynamic irons) are designated by the letter symbol Э and two-digit or three-digit numbers: the first digit represents the approximate silicon content in per cent; the second digit, the conventional designation of guaranteed magnetic properties; the third digit (which can be only zero) indicates that a given steel belongs to the cold-rolled class. The absence of the third digit indicates that the steel has been hot-rolled. For instance, grade Э41 is hot-rolled electrical steel containing about 4 per cent silicon (transformer iron) with normal specific core losses when remagnetised at 50 cycles.

As compared with carbon structural steels, alloy structural steels are characterised by higher tensile strength and yield point (Table 4). The latter property is of prime importance, since the higher values of yield strength indicate higher values of permissible (rated) stresses.

Table 4

**Comparison Between the Mechanical Properties of Carbon
and Alloy Structural Steels in Heat-Treated Condition
(hardening and high tempering)**

Grade of steel	Tensile strength	Yield point	Elongation	Contraction	Impact value
	not less than				
40	60	34	19	45	9
45	64	36	17	40	8
40X	80	65	12	40	7
40XC	90	70	12	45	6
40XH	83	65	13	45	8
OX113M (0.35% C)	100	84	13	45	8
OX114M (0.35% C)	105	90	12	40	8

Among the structural steels is especially singled out the group of steels used for springs; the chemical composition and mechanical properties of these steels are specified by the U.S.S.R. State Standard. Carbon spring steels are characterised by high carbon content ranging from 0.60 to 0.90 per cent, whereas in alloy spring steels the carbon content ranges from 0.45 to 0.70 per cent.

High speed steels hold a unique place among alloy tool steels. The remarkable property of these steels is their high *mechanical strength* at elevated temperatures (heat resistance). By this property is understood the power of hardened steel to retain its hardness when heated to elevated temperatures. The heating of a cutting tool during use causes self-tempering of the steel with a subsequent loss of hardness. Tools of hardened carbon steels usually retain their hardness up to 200°C. When heated above this temperature they begin to soften rapidly, losing their cutting properties and becoming "blunted", so that the further machining of metal becomes impossible.

The high-temperature strength of alloy tool steels is somewhat higher. High-speed steels retain their hardness even at temperatures of 600° C. A temperature of 600° C is that of low red heat, so the high heat resistance of high-speed steels is usually called a "red-hardness". High-speed steels are alloyed with tungsten, chromium and vanadium. The U.S.S.R. State Standard specifies two grades of high speed steels—P18 (containing 18 per cent tungsten), and P9 (containing 9 per cent tungsten).

High-alloy steels *with special physico-chemical properties* include: stainless, acid-resisting, non-scaling, heat-resisting steels, steels with high ohmic resistance, electrical, magnetic, non-magnetic, and wear-resisting steels.

Stainless steels possess a high resistance to atmospheric corrosion and to attack by humidity at ordinary and elevated temperatures. All types of stainless steel are alloyed with chromium. The main grades of stainless steel are as follows: 1X13, 2X13, 3X13 and 4X13. The stainless steels are used for steam turbine blades, valves for hydraulic pumps, household articles and utensils (knives, forks, washing-machine tanks, etc.).

Acid-resisting steels possess high corrosion resistance to attacks by various acids and other chemicals. Like stainless steels, all acid-resisting steels are alloyed with chromium. Most acid-resisting steels also contain nickel, and to some of them are added manganese, molybdenum, titanium and niobium. Titanium and niobium are added specifically to improve the weldability of these steels. When acid-resisting steels which contain either titanium or niobium are welded, carbides form, on account of which areas of metal near to the weld seam lose to a considerable extent their corrosion resistance, so that they rapidly corrode in service. But when acid-resisting steels, containing titanium or niobium, are welded, no structural transformations occur in the weld area, which retains its high corrosion resistance. The most widely used acid-resisting steels are: X17, X25, X28, 1X18H9T, 2X18H9, X18H11B, X18H12M3T.

Non-scaling steels (which are sometimes wrongly called heat-resisting) have high resistance to gas corrosion at elevated temperatures. All non-scaling steels contain chromium, and some are also alloyed with silicon, aluminium, nickel and titanium. Steels of grades X18H25C2 and X25H2OC2 are not liable to scaling when heated to about 1,000 to 1,100° C. The non-scaling steels are used for parts of furnace conveyers, muffles, thermocouple protective tubes, etc.

Non-scaling steels that retain much of their strength at elevated temperatures are called *heat-resisting* steels. All heat-resisting steels are alloyed with chromium, and they may also contain molybdenum (X5M), molybdenum and silicon (X6CM, X7CM, X10C2M), nickel and silicon (X13H7C2), nickel, tungsten and molybdenum (4X14H14B2M, 1X14H14B2M), nickel, molybdenum and titanium (X18H12M3T), nickel, silicon, tungsten and molybdenum (X14H14CB2M). The acid-resisting chromium-nickel steel 1X18H9T alloyed with titanium is also classed as a heat-resisting steel. The less alloyed steels, in particular the so-called "silchroms" (steels, alloyed with chromium, silicon and molybdenum, of grades X6CM, X7CM, X10C2M), are used for parts employed at about 350 to 500° C. They are also used for valves of car engines. The higher

alloyed steels are employed at higher temperatures (500 to 650° C). These steels are suitable for parts of gas turbines and jet engines. The nickel- or cobalt-base alloys are used at even higher temperatures (650 to 800° C).

Steels of *high ohmic resistance* possess high electrical resistivity, attaining 1.0 to 1.3 ohm $\times$ mm² $\times$ m, i. e., 10 to 12 times that of iron. At the same time they are non-scaling. Therefore these steels are used for wire and ribbons, which are employed for electric furnace heaters. All these steels are alloyed both with chromium and aluminium. They are also characterised by very low carbon content (less than 0.1 per cent). The less the carbon content in such steels, the higher the temperature they can withstand for long period in service.

Electrical steels are divided into two groups: *dynamic* and *transformer* irons. Electrical steels are alloyed with silicon, whose content in dynamic irons of grades Э1, Э2 and Э3 is approximately equal to 1, 2 and 3 per cent respectively, and in transformer iron of grade Э4, about 4 per cent (the U.S.S.R. State Standard does not specify the precise chemical composition of these steels, since grades of electrical steels are distinguished according to their magnetic properties). In electrical steels the carbon content should not exceed 0.1 per cent, since it greatly impairs the magnetic properties of these steels. Electrical steels are characterised by high magnetic induction, extremely low coercive force and low watt losses (i. e., low heat losses by electric power required for remagnetising). Dynamic steel sheets are used for stamping stock for poles and other magnetic conductors of electrical machines, and sheets of transformer iron are used for transformer cores.

Unlike the electrical steels, those used for *permanent magnets* should, in addition to high magnetic permeability, possess high coercive force. Steels for permanent magnets are alloyed with chromium, tungsten and cobalt. The U.S.S.R. State Standard specifies five grades of steels for permanent magnets, two chromium (EX and EX3), one tungsten (E7B6), one chromium-cobalt (EX5K5), and one chromium-cobalt-molybdenum (EX9K15M). The cobalt steels possess especially high magnetic properties. Recently magnet steels have been increasingly replaced by cast iron-nickel-aluminium alloys of types such as Alni, Alnisi, Alnico and Magnico.

Non-magnet (more precisely, *low-magnet*) steels are used for those parts of electrical machines that must be as strong as steel and as non-magnetic as copper or bronze, e.g., bandage rings for turbogenerators. Among the non-magnet steels, grade H9Г9, containing 0.50 per cent carbon, is most widely used.

Wear-resisting steels are especially advantageous because of their resistance to wear, i. e., resistance to abrasion by sand, earth, rock, ore and other hard substances, which is accompanied by cold harden-

ing of steel. Wear-resisting steels are used for jaw plates and other parts of rock crushing machines, cases of edge and ball mills, dipper teeth for power shovels and excavators, railroad crossings, frogs, switch points, guard rails, etc. Among the wear-resisting steels only one grade may be mentioned, i. e., high-manganese steel, occasionally referred to as Hadfield steel (after the name of prominent English metallographer R. A. Hadfield who first developed it). This steel contains 1.0 to 1.3 per cent carbon and 11 to 13 per cent manganese. Steel of this grade is usually designated as $\Gamma 13$.

4. THE IRON-CARBON PHASE DIAGRAM

The diagram in Fig. 1 gives the iron-carbon alloy system. This diagram refers to conditions of iron-carbon alloys, when, in the course of their primary and secondary crystallisations, carbon separates from liquid and solid solutions not in the pure form (as a graphite), but as a chemical compound, i. e., as an iron carbide or cementite (Fe_3C). Consequently this constitutional diagram (or phase diagram) covers only steels and white irons. For the study of gray irons this diagram alone is insufficient.

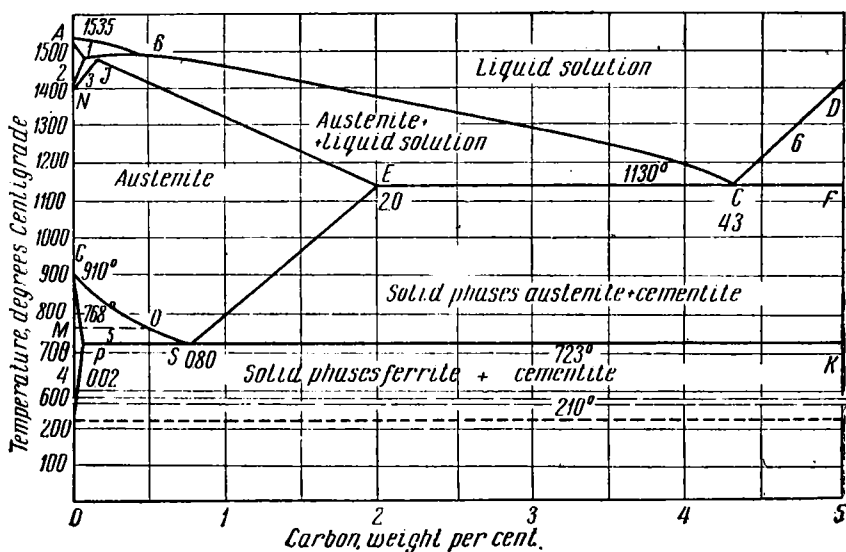


Fig 1. Iron-carbon phase diagram.

1—liquid solution + solid solution; 2—solid solution; 3—solid solution + austenite; 4—solid solution (ferrite); 5—ferrite + austenite; 6—liquid solution + cementite.

In several areas of the diagram there are indicated names of the phases existing at temperatures and concentrations determined by the boundaries (lines) of these areas.

It should be noted that by the term "*a phase*" is meant a *perfectly homogeneous (in physical and chemical respects) portion of an alloy (with one common pattern of space lattice) separated from the remainder by distinct bounding surfaces*. A liquid solution and a solid solution of the same chemical compositions constitute different phases, because their respective physical, or aggregate states are different. The two allotropes of iron, alpha iron and gamma iron, are also different phases because their space lattices differ.

The portions of the diagram which lie between the *liquidus* and *solidus* lines (*ABCD* and *AIECF* respectively) represent the process of *primary* crystallisation on freezing the alloys, i. e., the solidification of the liquid solution, whereas those areas between lines *GSECF* and *PSK* represent the process of *secondary* crystallisation, i. e., decomposition of solid solution or austenite.

Most of the heat-treating operations applied to steel (some kinds of annealing, normalising, hardening, carburising, etc.) are connected with the processes of secondary crystallisation. We shall therefore consider more closely the left-hand bottom portion of the iron-carbon phase diagram (Fig. 2).

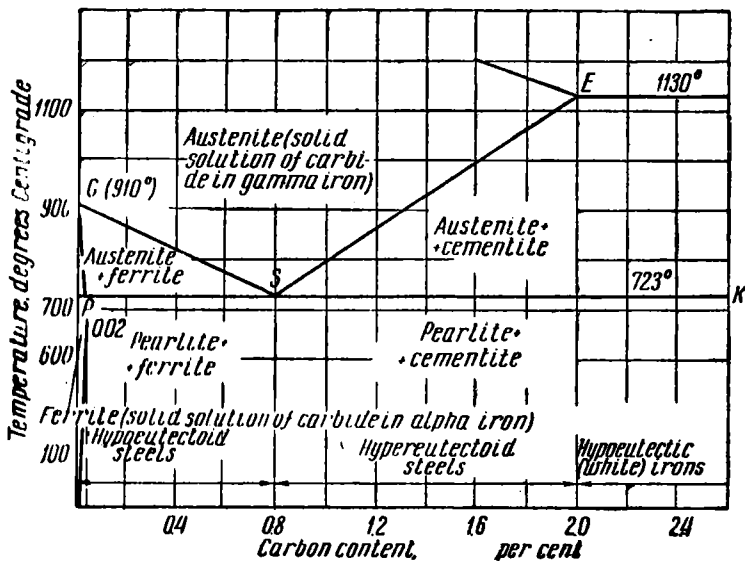


Fig. 2. Left hand bottom ("steel") portion of the iron-carbon phase diagram (area of the secondary crystallisation of steel).

At temperatures above lines *GS* and *SE* only one phase, *austenite*, exists.

Each point on the lines *GS* and *SE* represents temperature, at which the process of austenite decomposition begins in freezing steel. In hypoeutectoid steels containing up to 0.8 per cent carbon, decomposition of austenite takes the form of the separation from it of a new phase, called *ferrite*. Separation of ferrite takes place continuously during the decomposition of cooled austenite, beginning at temperature A_1 (on line *GS*) and reaching completion at temperature A_1 (on line *PSK*). As the ferrite separates, the carbon concentration in the austenite progressively increases, reaching the value of 0.8 per cent (the *S* point) at a temperature of 723° C (line *PSK*).

In hypereutectoid steels containing more than 0.8 per cent of carbon the decomposition of austenite takes the form of the separation from it of a *cementite*. The separation of cementite takes place continuously during the decomposition of cooled austenite, starting at a temperature A_1 (or the A_{cm} point on line *ES*) and being completed at a temperature A_1 (on the *PSK* line). As the cementite separates, the carbon concentration in the austenite continuously decreases, also reaching the value of 0.8 per cent (the *S* point) at a temperature A_1 (on line *PSK*).

Thus, in slow cooling any steel, hypoeutectoid as well as hypereutectoid, has in its structure austenitic grains containing 0.8 per cent carbon as well as ferrite or cementite. At the A_1 point, the process of secondary crystallisation ends with the *eutectoid decomposition* of the remaining austenitic grains. The austenite becomes a conglomerate consisting of two phases, ferrite and cementite.

The structure of steel is formed as a result of these transformations. However, the knowledge of the iron-carbon phase diagram only is insufficient to predetermine precisely what this structure will actually be. This will be explained by the following example.

As already indicated, at the A_1 temperature the secondary crystallisation of steel ends with the eutectoid decomposition of austenite into a conglomerate consisting of two phases, ferrite and cementite. On relatively slow cooling, this conglomerate of ferrite and cementite becomes sufficiently distinctive for its structure to be well enough examined at a magnification of about 500 diameters. Such ferrite-cementite mixture is known as *pearlite*. With more rapid cooling the austenite is also transformed into a mixture of ferrite and cementite, but the structure of this conglomerate is less distinctive, so that in order to examine it substantially greater magnifications (about 1,000 to 1,500 diameters) must be used. Such a mechanical mixture of ferrite and cementite is called *sorbite*. With still more rapid cooling rates a ferrite-cementite conglomerate will

also be obtained, but the structure of this modification of pearlite is so fine that it can be distinguished only by means of an electron microscope. Such an aggregate of ferrite and cementite is called *troostite*.

Thus, in all three cases the phase composition of steel is of exactly the same pattern, i. e., ferrite and cementite, but in each case the structure of the steel and its properties will vary markedly.

From this example an extremely important conclusion may be drawn, i. e., that the *formation of the structure of steel* (or of any other alloy) *is determined by the kinetics of phase transformations*, i. e., by the process of this transformation with time. The practice of heat-treating steel is based to a considerable extent, if not entirely, on a consideration of the kinetics of phase transformations that take place during the secondary crystallisation of steel.

It is also very important to note that under real conditions the transformation temperatures A_1 and A_2 differ from those indicated in the phase diagram. Under real conditions the austenite transforms into a ferrite cementite conglomerate not at 723°C , as is indicated in the diagram, but at a slightly *lower* temperature. The reverse change of the ferrite cementite aggregate to austenite on heating takes place at temperatures higher than 723°C , i. e.,

$$A_1 \geq 723^\circ \text{ on heating,}$$

$$\text{and } A_1 \leq 723^\circ \text{ on cooling.}$$

The real value of the A_1 temperature depends upon the speed of heating or cooling.

Thus, it is evident that, firstly, the iron-carbon phase diagram cannot explain the process of formation of a specific steel structure under real conditions of heat-treatment, and secondly, the real temperatures at which phase transformations occur under the same conditions differ in a greater or lesser degree from those indicated in the phase diagram. The question naturally arises as to what the iron-carbon phase diagram can indicate in such a case, and what help it can afford in the practical accomplishment of heat-treating operations.

The answer is as follows: firstly, on the basis of this diagram it can be quite accurately ascertained what phase composition will have an alloy in a condition of equilibrium at any temperature, and consequently, what phase composition is likely to arise in an alloy being in a state of non-equilibrium. The equilibrium condition of a system determines the direction of spontaneous processes that take place or may take place in a system which is not in a state of equilibrium.

Secondly, a sufficient approximation of an alloy to a condition of phase equilibrium is reached on its being slowly heated to high temperatures. So the actual temperatures of phase transformations closely coincide with those indicated by the diagram. Therefore the selection of the temperatures of heating for annealing, normalising and hardening is made on the basis of the iron-carbon phase diagram.

Thirdly, this diagram facilitates the establishing of the temperature ranges of phase transformations.

A condition of steels approaching equilibrium is attained on heating very slowly from temperatures above the A_1 point and is accomplished practically by annealing. Depending upon the carbon content, the structure of annealed carbon steels at room temperatures will be composed of:

pearlite and ferrite: hypoeutectoid steels (less than 0.8 per cent carbon);

pearlite: eutectoid steel (0.8 per cent carbon);

pearlite and cementite: hypereutectoid steels (more than 0.8 per cent carbon).

As to terminology, there must be made one reservation. It is often said that "the structure of steel is composed of pearlite and ferrite". This is not altogether correct: each steel, as well as each crystalline body, consists of grains (crystals). Thus, it is more exact to say that "the structure of steel is composed of pearlite and ferrite grains". However, if it is stipulated that by ferrite is meant not only a certain phase but an aggregate of all ferrite grains, and under the term pearlite not only the structural constituent but an aggregate of all pearlite grains, the above definition may be further abbreviated in the following manner: "The structure of steel consists of pearlite and ferrite."

5. PHASE TRANSFORMATIONS IN ALLOY STEELS

The transformations occurring under conditions of equilibrium in steels alloyed with one element only should be studied by means of ternary diagrams, i. e., the iron-carbon-alloying element, while those in steels alloyed with two elements should be investigated by means of quaternary diagrams, etc. But the ternary diagrams are very complicated and not yet sufficiently studied, so that their practical usage involves considerable difficulties. Still more complex are the quaternary diagrams. Therefore in the study of alloy steels binary phase diagrams of the iron-alloying element type are usually considered; in this way is studied the influence of alloying elements

on the positions of lines and transformation points in the iron-carbon constitutional diagram.

With regard to chief alloying elements, all phase diagrams of type iron-alloying elements may be reduced to two main types (Fig. 3). They differ from each other mainly in the manner in which

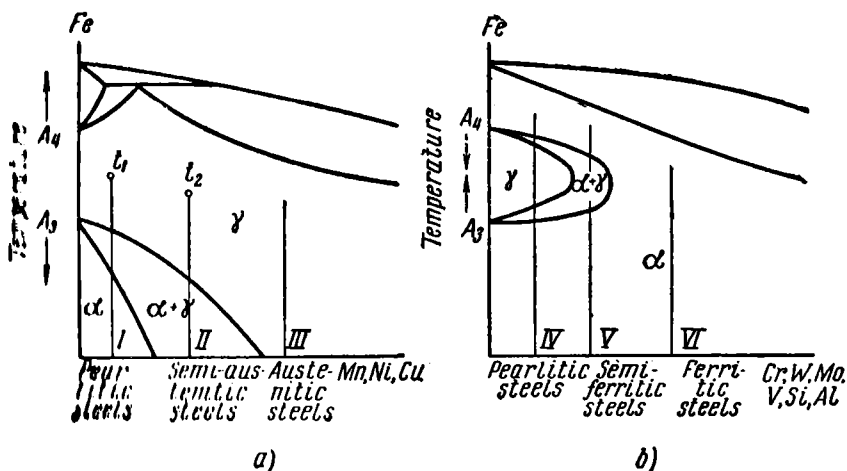


Fig. 3. Patterns of the main phase diagrams of the iron-alloying element type.

the alloying elements influence the widening or narrowing of the area where the gamma solution, i. e., solid solution of alloying element in gamma iron, exists.

Some alloying elements (Ni, Mn) broaden the area of gamma iron (Fig. 3, a) and hence depress the A_1 point and raise the A_2 point. The greater the content of alloying element in an alloy, the higher the temperature at which the transformation A_1 takes place and the lower the temperature at which the transformation A_2 occurs. From this it follows that alloys low in alloying element (alloy I) consist (at room temperature) of a solid solution of alloying element in alpha iron, called an alpha solution. As the temperature is raised complete phase recrystallisation occurs, and at a temperature t_1 the alloy I consists of gamma solution.

At room temperature alloys with larger concentrations of alloying element (alloy II) are made up of two phases, alpha solution and gamma solution. As the temperature rises the alpha solution changes to gamma solution, and at the temperature t_2 the alloy II consists of the gamma solution only. Thus, in alloys like alloy II only partial phase recrystallisation takes place.

Finally, with a still greater content of alloying element (alloy III) the alloy consists of gamma solution only at all temperatures. Consequently in alloys of type III the phase recrystallisation does not occur at all.

The presence of carbon in these alloys, turning them into alloyed steels, produces in general no effect whatsoever upon the character of the phase transformation process described, though it gives rise to some quantitative changes.

Low alloy steels (like alloys of type I) at room temperature are made up of two phases—ferrite (alpha solution) and carbide (cementite if the alloying element does not form carbides, or complex carbide of type Fe_nMe_nC if the alloying element is carbide forming). According to the carbon content, these steels in annealed condition at room temperature may be classed as hypoeutectoid (ferrite and pearlite) and hypereutectoid (pearlite and secondary carbides). Complete phase recrystallisation, similar to that taking place in carbon steels, occurs in these steels at temperatures A_1 – A_c (Fig. 1). These steels are known as *pearlitic*.

At room temperature the structure of the higher alloyed steels (corresponding to alloys of type II) in an annealed condition is composed of austenite and pearlite (plus ferrite and secondary carbides, depending upon the carbon content). At a temperature range A_1 – A_c (Fig. 1) partial phase recrystallisation occurs: pearlite+ferrite, pearlite, pearlite+secondary carbides transform into austenite. The austenite phase, which existed in structure at room temperature, does not take part in the phase recrystallisation process. These steels are known as *semi-austenitic*.

Highly alloyed steels (corresponding to alloys of type III) have the austenitic structure at all temperatures. The phase recrystallisation does not take place in these steels. They are classed as the *austenitic steels*.

Other alloying elements (chromium, tungsten, molybdenum, vanadium, silicon, aluminium) tend to narrow the area in which the gamma solution exists (Fig. 3, b). At room temperature all alloys of iron with the alloying elements of this group are composed of alpha solution only. Alloys IV (with low content of alloying element), V (with higher content of alloying element) and VI (with high content of alloying element) differ from each other in their tendency to phase recrystallisation, i. e., alloys IV are capable of complete phase recrystallisation, in alloys V the phase recrystallisation covers only part of the alpha solution, and in alloys VI it does not occur at all.

Steels slightly alloyed with elements of this group (corresponding to alloys IV) in annealed condition and at room temperature are composed of pearlitic grains and, according to carbon content, of

ferritic grains and secondary carbides. These steels are capable of complete phase recrystallisation. They are known as pearlitic steels.

Only partial recrystallisation can occur in steels with a higher content of alloying elements (corresponding to alloys V). These steels are classed as *semi-ferritic*.

Finally, in highly alloyed steels (corresponding to alloys VI) phase transformations do not take place at all. At all temperatures their structure is composed of alpha solution or, if the steel is alloyed with a carbide-forming element and the amount of carbon present is substantial, of alpha solution and carbides. These steels are classed as *ferritic*. The ledeburite eutectic is found in the structure of high-carbon steels alloyed both with the elements of the first and second groups. As the result of forging the ledeburite eutectic is broken down into a series of isolated ledeburite carbides. These steels are identified as steels belonging to the carbide class.

Thus, alloy steels may be divided into six structural classes in accordance with their ability to undergo phase recrystallisation and their phase composition at room temperature. If the practically irrelevant semi-austenitic and semi-ferritic classes are excluded, most alloy steels may be divided into four main structural classes—pearlitic, austenitic, ferritic and carbide.

All low-alloy structural steels, the greater part of alloy tool steels (with a relatively low carbon content), and stainless steels of grades 1X13-4X13 are *pearlitic*. The structure of a steel of this pearlitic class in annealed condition is represented by Fig. 4.

The highly and complexly alloyed acid-resisting steels of grades 0X18H9, 1X18H9, 1X18H9T, 2X18H9 and X13H4Г9, non-magnetic steels and wear-resisting steel of grade П13 are classed as *austenitic*. The structure of the steel of austenitic class is shown in Fig. 5. The fact that the most typical acid-resisting steels are specified as austenitic is explained by the following considerations; as is well known from the theory of corrosion, alloys with a homogeneous (single phase) structure are more resistant to corrosion than those with a heterogeneous structure. It is clear, too, why non-magnetic steels should have the austenitic structure, since among all the phases of iron-carbon alloys only austenite is non-magnetic. The high wear-resisting properties of articles made of high-manganese steel П13 are explained as follows: the austenite transforms into martensite by abrasive grinding accompanied by cold hardening, so that the surface hardness of the steel increases greatly.

The *ferritic* class includes the highly alloyed chromium non-scaling steels of grades X25, X25T, X28, high-ohmic resistance steels,

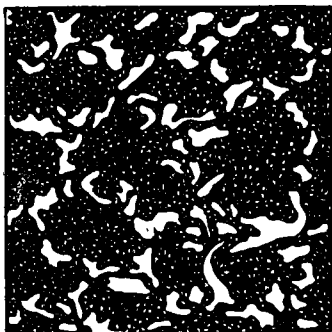


Fig. 4. Structure of alloyed steel of the pearlitic class in annealed condition. Ferrite and sorbite, 300 $\times$.

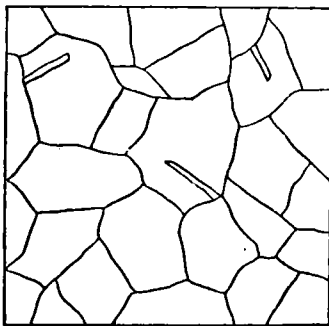


Fig. 5. Structure of alloyed steel of the austenitic class. Austenite, 300 $\times$.

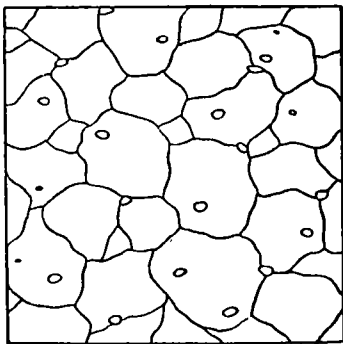


Fig. 6. Structure of alloyed steel of the ferritic class. Ferrite and carbide, 300 $\times$.

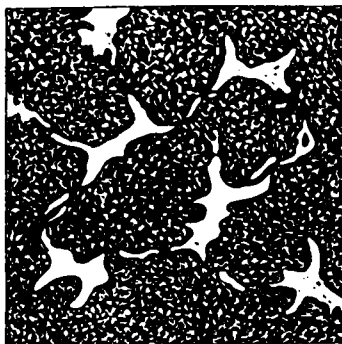


Fig. 7. Structure of high-speed steel in as-cast condition, 500 $\times$.

and transformer irons. The structure of steels of the ferritic class is shown in Fig. 6.

The *carbide* class comprises some tool steels, such as chromium steels of grades X12 and X12M, and high-speed steels of grades P9 and P18. The structure of steels of the carbide class in as-cast condition contains the ledeburite eutectic (Fig. 7). Forging and rolling cause the decomposition of the ledeburite eutectic into separate grains, the structure of forged steels of the carbide class in an annealed condition being composed of grained pearlite and of larger grains of secondary and ledeburite carbides (Fig. 8).

Now let us consider the effect of alloying elements on the positioning of transformation points and lines in the phase diagram for the iron-carbon alloys.

All alloying elements with the exception of aluminium and cobalt displace the transformation points S and E in the diagram to the left, this effect increasing with the carbon content in the steel. This may account, in particular, for the fact that high speed steels containing

only 0.70 to 0.95 per cent carbon have been found to belong to the carbide class, i. e., to lie to the right of point E in the phase diagram. As for the value of temperature A_1 , it follows from the diagrams in Fig. 3 that the alloying elements causing the narrowing of the gamma-phase region (tungsten, molybdenum, vanadium, mangan, aluminium) raise the A_1 point, whereas those elements widening the gamma region (nickel, manganese) lower this point. As to the effect of chromium, it raises the A_1 point when contained in large amount, and depresses it when its proportion is small. The effect of alloying elements on the position of the A_1 point is similar.

From what has been said it is evident that the hardening temperature for chromium, tungsten and other steels alloyed with elements that narrow the gamma-phase region should be higher than that for carbon steels of the same carbon content, whereas that for nickel and manganese steels should be lower.

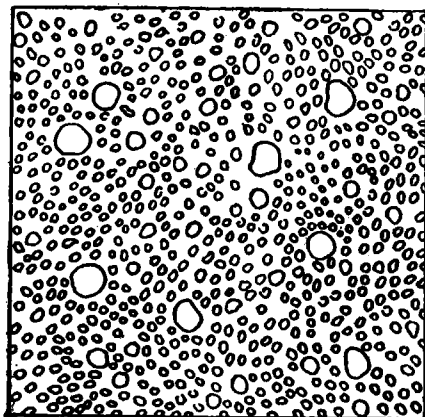


Fig. 8. Structure of alloyed steel of the carbide class (high-speed steel of grade P18) in annealed condition. Pearlite grains and ledeburite and secondary carbides are visible, 500 $\times$.

Chapter II

TRANSFORMATIONS IN STEEL ON HEATING

6. TRANSFORMATION OF PEARLITE INTO AUSTENITE

The transformation of the two-phase conglomerate of ferrite and cementite, known as pearlite, into the solid solution, or austenite, takes place at the A_1 temperature. For carbon steels the A_1 temperature on very slow heating is 723°C . Under real conditions of heating the pearlite transforms into austenite at higher temperatures ($A_1 > 723^\circ\text{C}$).

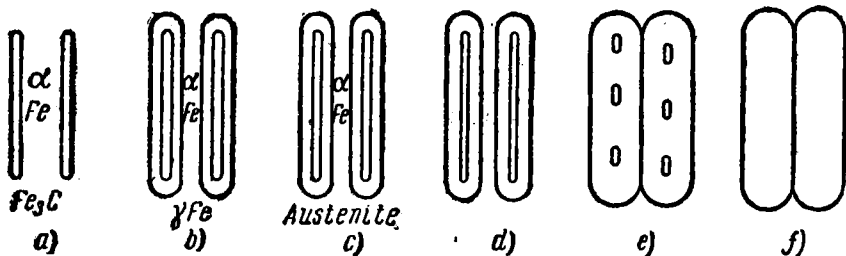


Fig. 9. Diagram illustrating the transformation of pearlite into austenite.

Now let us consider how the transformation of pearlite into austenite will proceed in steel of eutectoid composition. For this purpose a part of a pearlite grain will be examined (Fig. 9, a). The cementite lamellas, or plates, are in direct contact with ferrite. The carbon atoms (contained in cementite) always tend to penetrate into the space lattice of the iron, this tendency becoming more intensive with an increase of temperature. In fact the solubility of carbon in ferrite, though insignificant in its absolute value, increases gradually with the rise of temperature (see Fig. 2).

On reaching the A_1 temperature the space lattice of alpha iron changes to that of gamma iron in the areas immediately adjoining the cementite lamellas (Fig. 9, b). Subsequently there takes place a process of gradual dissolution of the cementite in these areas of gamma iron just formed (Fig. 9, c). This results in the formation of the primary austenitic grain in which the surrounding ferrite dissolves, giving rise to a gradual growth of austenitic grains. At the same time a dissolving of the cementite lamellas in these grains takes place (Fig. 9, d). Both these processes continue until the complete dissolving of ferrite and cementite is achieved. According to experimental data, the dissolving of the ferrite is completed before that

of the cementite. In other words, at the moment when the dissolving of ferrite comes to an end, there are still small grains of pearlitic cementite remaining in the steel structure (Fig. 9, e). When these grains of cementite are dissolved, the steel structure becomes homogeneous in so far as its phase composition is concerned and is composed of austenitic grains only. But the grains of austenite remain chemically heterogeneous for some time, which is explained as follows: the concentration of carbon is high in those areas where cementite lamellas existed, while in peripheral regions it is low (Fig. 9, f). The carbon concentration is gradually levelled by diffusion in the whole austenitic grain. Chemically homogeneous grains of austenite are thus obtained.

From a consideration of this transformation pattern the conclusion may be drawn that an independent austenitic grain forms near to each cementite lamella. But as soon as these primary austenitic grains touch each other the process of their coalescence begins, through collective recrystallisation. This process proceeds simultaneously with the last stages of the transformation process. When the transformation is completed, there remain a number of austenitic grains instead of one grain of pearlite.

Such is the *mechanism* of the transformation of pearlite into austenite. We shall now discuss the *kinetics* of the transformation, i. e., the proceeding of the transformation process with time.

The kinetics of the transformation of pearlite into austenite is studied in the following manner. Small steel specimens are rapidly immersed into a bath heated to some constant temperature above that corresponding to equilibrium. Since the specimens are small, it can be accepted that they are very rapidly heated up to the bath temperature.

The aim of the experiment is to determine the speed of the transformation of pearlite into austenite. Readings are taken, beginning from the moment the specimen is put into the furnace. The degree of transformation can be estimated in several ways. For instance, a number of completely similar specimens may be immersed in a bath and then removed one by one at equal time intervals in order to be immediately quenched; their structure can then be examined and their hardness determined. Obviously the higher the degree of transformation, i. e., the larger the amount of pearlite transformed into austenite, the greater the quantity of martensite in quenched specimens and the higher their hardness. Thus, it is possible to observe the proceeding of the transformation process with time by the changes of structure and hardness or some other physical property.

The degree of the transformation of pearlite into austenite can be estimated in a still simpler manner, i. e., by the changes of the

magnetic properties of specimens during the transformation process. Pearlite is magnetic, and austenite non-magnetic; as the transformation proceeds, the magnetic properties begin to decrease gradually until they reach zero on the completion of the process.

To record the changes of magnetic properties during transformation special devices such as magnetometers are used. A schematic

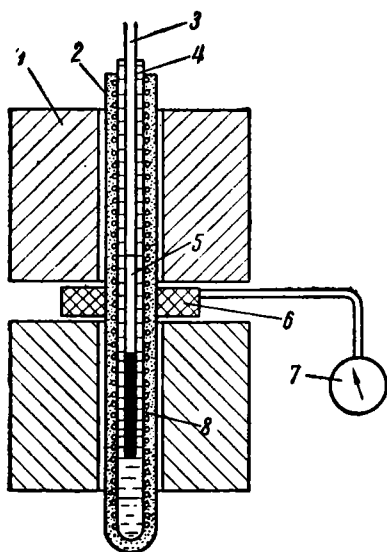


Fig. 10. Schematic diagram of a magnetometer.

diagram for one of them is shown in Fig. 10. The magnetometer consists of a powerful permanent magnet 1 which is drilled through axially, a tube 4 filled with molten salt being inserted into the hole. The tube is wound round with nichrome wire which is protected with heat-insulating matter 2. So tube 4 represents a bath which can be heated up to a certain given temperature. A specimen to be tested, 8, is fixed at one end of rod 3 which is made of non-magnetic metal. The phase transformations occur in the specimen 8, immersed in the bath. The specimen is raised periodically to an interpolar space, position 5. This changes the intensity of magnetic flux, recorded by a measuring ballistic coil 6 connected with the galvanometer 7. The advantage of this method is that a single specimen

is sufficient to illustrate the transformation process from beginning to end. Hardening is no longer necessary.

Fig. 11 gives the results obtained in studying the kinetics of transformation of pearlite into austenite at several constant temperatures (i. e., the kinetics of the isothermal transformation). From this diagram it follows that the lower the temperature of isothermal heating, the longer is the transformation process.

From the data of this diagram another more convenient one can be drawn with time and temperature plotted as its coordinates (Fig. 12). The solid line in this diagram indicates the temperatures of the complete transformation of pearlite into austenite, if at constant temperature (isothermally). It is probable that the transformation process begins as soon as the steel has reached a temperature of isothermal heating. So if the steel is put inside a furnace with temperature t_1 and has reached this temperature, the beginning of the transformation of pearlite into austenite is determined by the

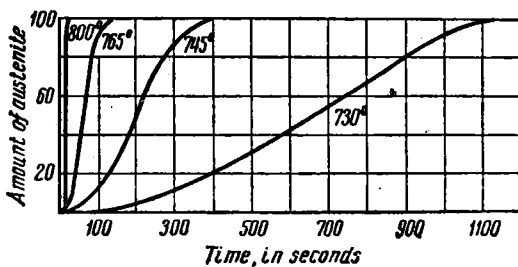


Fig. 11. Kinetics of the transformation of pearlite into austenite.

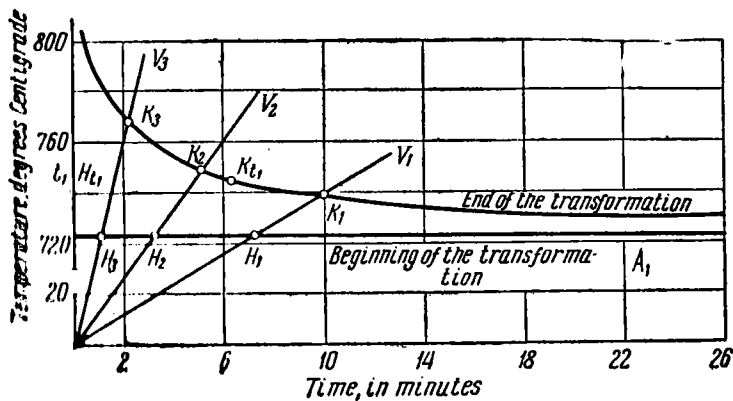


Fig. 12. Summary kinetic diagram of the transformation of pearlite into austenite.

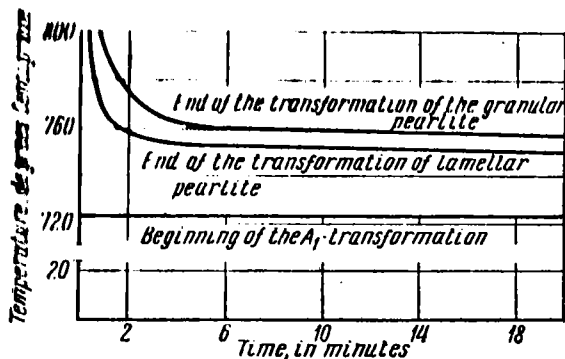


Fig. 13. Effect of the form of pearlite on the kinetics of its transformation into austenite.

point H_{t_1} and the end by the point K_{t_1} . The duration of the transformation is determined by the section $H_{t_1}-K_{t_1}$.

From this diagram (Fig. 12) the kinetics of the transformation on continuous heating may be determined. There are three straight lines v_1 , v_2 and v_3 , which represent the heating operations at different heat rates; the steeper the line, the greater the heating rate. Thus, the lines v_1 , v_2 and v_3 correspond to the low-rate, high-rate, and extremely high-rate heating, respectively. The transformation under continuous heating at any rate occurs at the same temperature, which is equal, or more precisely, somewhat higher than the A_1 temperature (the points H_1 , H_2 , H_3). The end of the transformation process is determined by the points K_1 , K_2 and K_3 .

It is therefore clear from this diagram that the transformation of pearlite into austenite on continuous heating occurs through a temperature range. The slower the heating, the narrower the temperature range for the transformation process, and on indefinitely slow cooling it is reduced to zero.

The diagram represents the isothermal transformation of lamellar pearlite into austenite. The granular pearlite transforms at a considerably lower rate (Fig. 13); according to the data of S. S. Steinberg the isothermal transformation of granular pearlite into austenite at 750° C takes five times longer than the transformation of lamellar pearlite at the same temperature.

7. LAWS GOVERNING THE GROWTH OF AUSTENITIC GRAINS

At any temperatures from above the A_1 range up to the melting point, the structure of any steel is composed of austenitic grains. At temperatures above the A_1 range no phase transformations take place, but structural changes occur, i. e., the grains of austenite grow.

The interesting feature is not the growth of austenitic grains in itself but the fact that the grains of austenite in various steels possess different tendencies to growth. In steels of one kind the grains grow very rapidly (Fig. 14) and become very large after heating up to the temperature of 930° C. Such steels are called originally or inherently coarse-grained. In other steels the austenitic grains grow very slowly, and the steel retains the fine grain structure at the same temperature of 930° C. Such steels are said to have an original or inherent fine grain structure.

To estimate the tendency of austenitic grains to growth a temperature of 930° C is usually selected, being in practice the highest to which most steels are heated during heat-treatment. But this method of estimation, though it is quite adequate from the practical point of

view, is very conventional, as is evident from diagram (Fig. 14). On further heating above the temperature 930°C the austenitic grains of fine grain steel begin to grow rapidly, catching up and even overtaking in this respect the coarse grain steels.

What is the cause of the different tendencies to growth of the austenite in various kinds of steel?

There is as yet no established opinion, but the most plausible explanation is as follows: the growth of austenitic grains is hampered by the finest ultramicroscopical particles of nitrides, oxides, and carbides distributed at the grain boundaries of austenite and forming as it were "barriers" between the grains. In fact it has been established that steels deoxidised by aluminium or modified by boron, titanium and other similar elements always possess inherent fine grain structure. But as the steel is heated these particles of nitrides, carbides, and oxides gradually dissolve in the austenite or coagulate (coalesce), the so-called "barriers" disappear and nothing remains to impede the growth of austenitic grains.

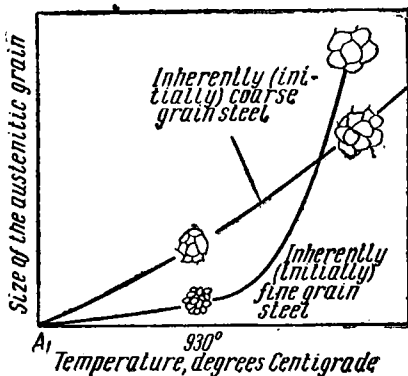


Fig. 14. Diagram illustrating the growth of austenitic grains.

6 EFFECT OF AUSTENITIC GRAIN SIZE ON THE PROPERTIES OF STEEL

The size of austenitic grains exerts a great immediate or indirect influence upon many properties of steel.

The austenitic grain size cannot produce any direct effect on the mechanical properties of steel, since the grains of austenite do not exist in steel structure at room temperatures, at which the mechanical properties are usually determined (this obviously does not apply to steels of the austenitic class). But an indirect influence of the austenitic grain size upon the mechanical properties of steel is unmistakable, i. e., the larger the austenitic grains, the larger under otherwise equal conditions will be the grains of ferrite and pearlite, the mechanical properties, particularly impact strength, being directly dependent upon the size of these grains.

The direct effect of the austenitic grain size on the structural changes in steel during its hardening proves to be much more substantial.

The transformation of austenite into pearlite begins mainly at the surface of the austenitic grains. Thus, the finer the grains of austenite, the greater will be their summary surface area and consequently the more rapidly will proceed the process of their change to pearlite. From this it follows that in hardening steels with a fine grain austenite structure higher cooling rates should be employed in order to suppress the tendency to transformation into pearlite.

From what has been said it is evident that the depth of hardening penetration, or the so-called hardenability, for two steels with the same chemical composition but with different sizes of austenitic grains will be different on cooling at the same rate; the larger the size of austenitic grains, the greater the hardenability of the steel.

When steel with a structure of coarse-grained austenite is hardened, coarser acicular martensite develops, and consequently the internal stresses are greater; hence the extent of deformation is larger and the forming of quenching cracks more probable. On the other hand, in steels with a structure of small austenitic grains the slip planes preceding the forming of acicular martensite are considerably less pronounced, and this accounts for the decrease in the internal stresses, the degree of deformation and the probability of crack formation.

9. METHODS FOR DETERMINATION OF THE AUSTENITIC GRAIN SIZE

In the State Standards there are four standard methods for determining the austenitic grain size in steels of the pearlitic class:

- 1) carburising—for hypoeutectoid steels;
- 2) forming of carbide network—for hypereutectoid steels;
- 3) high-temperature etching in molten salt bath—for steels of any carbon content;
- 4) high-temperature oxidation—for steels of any carbon content.

In the *carburising* method the specimen of hypoeutectoid steel is carburised at a temperature of 920° C to 940° C for a period of 8 hours, the time of heating up to this temperature not being counted, and then slowly cooled down to 600° C, the rate of cooling not exceeding 100 degrees per hour for carbon steels or 50 degrees per hour for alloy steels. The rate of further cooling is of no importance. The austenitic grains develop on the carbide network of the carburised part of specimen.

In the method of determining the austenitic grain size by means of *carbide network* the specimen of hypereutectoid steel is heated up to the same temperature and is kept at this temperature for 3 hours, being cooled afterwards to 600° C at a rate of 80 to 100 degrees per hour.

In the method of *high-temperature etching* in molten salt bath a microsection of a steel specimen is heated in a furnace with protective gas atmosphere, or in a common furnace, being protected from the oxidation by the iron chips, borax, charcoal powder and waste carburiser, or in bath of molten sodium carbonate. The heating is accomplished at a temperature of 920° C to 940° C for a period of 3 hours, the hot-etching test being then conducted in a molten mixture consisting of equal quantities of barium chloride (BaCl_2), sodium chloride (NaCl) and calcium chloride (CaCl_2). The etching test is accomplished at a temperature of 930° C for 2 to 5 minutes, the cooling in kerosene being then carried out until all the salt is washed off. The austenitic grains usually become visible without the additional etching of a microsection.

In the *oxidation* method a microsection is heated in a furnace with an oxidising atmosphere at the same temperature for 3 hours and then is quenched in water. The austenitic grains become visible on the network of oxides after additional grinding, polishing and etching of microsection by a 15 per cent solution of hydrochloric acid for a period of 2 to 10 minutes.

The *ferrite network* may also be used as a means of determining the austenitic grain size in hypoeutectoid steels containing 0.3 to 0.5 per cent carbon.

The size of austenitic grain made visible by any one of the above-mentioned methods is determined through microscopic examination at a magnification of 100 diameters by comparing the structure of a microsection with the scales shown in the standard referred to. Altogether 42 grain numbers have been established:—1, 0, 1, 2, etc., up to 10 inclusive.

Vacuum metallography offers very interesting possibilities for the determination of austenitic grain size. Heating cross-sections of the steel specimens at high temperatures in a vacuum (10^{-4} to 10^{-5} mm of mercury column) produces a vapourisation of atoms from the surface of the cross-section which is the most intensive at the grain boundaries (selective evaporation), the boundaries of the austenitic grains being thus made visible. The austenitic grains distinguished by this method can be viewed by means of an ordinary microscope after cooling the cross-section.

10. CHANGES IN STEEL STRUCTURE ON HEATING NOT CAUSED BY PHASE TRANSFORMATIONS

When unhardened steel of the pearlitic class is heated at a temperature below the A_1 range, an increase of the solubility of carbon in ferrite is noticed. But as this increase is very insignificant

(0.01 per cent at room temperature and 0.02 per cent at 723° C), it cannot produce any marked changes in steel structure. However, on heating up to the A_1 temperature considerable structural changes in steel may occur. These changes can be of two kinds:

- 1) if the steel has been hardened by cold working, the recrystallisation process takes place on heating at a temperature below the A_1 range;

- 2) if the steel is kept for a long time at high temperatures (but below the A_1 point), the processes of coagulation and spheroidising of cementite occur.

The recrystallisation process consists in that at certain temperatures above the so-called recrystallisation point the refined and elongated grains in the steel structure, first gradually assume a regular geometrical (equiaxial) form (primary recrystallisation) and then grow larger as a result of the growth of some grains at the expense of the others (collective recrystallisation). The steel structure was made up of both pearlite and ferrite grains before the heating, as well as after it. Thus, its phase composition does not change during recrystallisation, whereas its structure does undergo the substantial changes. The properties of the steel change correspondingly, in that the tensile strength, yield point and hardness decrease, while those of elongation, reduction of area (characterising the ductility of the steel), and impact strength rose.

The spheroidising process consists in that the cementite plates tend to assume a globular form, and the coagulation process consists in the gradual disappearance of the small cementite grains and in the simultaneous coarsening of the larger cementite grains. The processes of spheroidising and coagulation (coalescence), proceeding simultaneously, are explained as follows. The small cementite grains are surrounded by the ferrite, which can dissolve carbon (though in small quantities). The solubility of carbon in the ferrite is not uniform: the carbon concentration is always higher in areas immediately adjoining small cementite grains than in those adjacent to larger ones. The solubility of different sides of cementite plates in ferrite is also not uniform. Therefore the carbon concentration in ferrite is unequal, and the presence of the concentration differential causes the diffusion of carbon, proceeding from those areas where its content is higher (i.e., near the small grains of cementite and narrow sides of the cementite plates) to those areas where its concentration is lower (i.e., near the larger cementite grains and broad sides of the cementite plates). The depletion in carbon of the first areas is made up by the gradual dissolution of small grains and the narrow sides of the cementite plates in ferrite, whereas the increase in carbon percentage of the ferrite areas near the larger cementite grains or broad sides of the cementite plates

gives rise to a separation of excess carbon in the form of the cementite molecules from the ferrite and to their precipitation on larger grains and the broad sides of the cementite plates. As a result of these processes the small grains and lamellas of cementite gradually disappear, while the larger grains progressively become coarser.

The processes of recrystallisation, spheroidising and coagulation are based on the common physical law known as the second principle of thermodynamics, i. e., each system (particularly each phase) spontaneously tends to diminish its store of free energy. This tendency is expressed in the reduction of the phase surface area. This is precisely the nature of the processes of recrystallisation, spheroidising, and coagulation, i. e., with primary recrystallisation and spheroidising the grains acquire an equiaxial (geometrically regular) form similar to the globular one, the sphere of ball possessing the smallest surface area possible of all bodies of the same volume; during the collective recrystallisation and coagulation (coalescence), the grains are coarsening, the surface area of one grain being always less than the sum-total of the surface areas of several grains having the same summary volume and of similar shape.

Chapter III

TRANSFORMATIONS IN STEEL ON COOLING

II. MECHANISM OF THE TRANSFORMATION OF AUSTENITE INTO THE FERRITE-CEMENTITE CONGLOMERATE

Austenite is a solid solution of the carbon, or of carbon and other elements in gamma iron. If steel is held at temperatures above the A_1 point for a sufficiently long period of time, the atoms of carbon are uniformly distributed in the space lattice of the gamma iron. But this uniformity is an average or statistical one. The carbon atoms may continuously move inside the space lattice, leaving some crystalline nuclei and penetrating into others. The average statistical uniformity of distribution of the carbon atoms in austenite is therefore continuously broken; in some areas of the austenite the proportion of the carbon atoms is greater than average, while in others it is less. Such breaches of uniformity of concentration are called *compositional fluctuations*. It is true that the regions with greater or lesser quantity of carbon atoms exist for only a very small time interval. The same continuous heat movement of carbon atoms that causes the fluctuation of the chemical composition tends to equalise the concentration of carbon atoms. Those portions of austenite grains that formerly possessed a high concentration of carbon

atoms subsequently become "normal" in this respect, whereas other portions of austenitic grains which were "normal", become momentarily supersaturated with carbon atoms; the third kind of grain areas, on the other hand, are depleted of them. This process goes on continuously and constantly. Among the portions of austenitic grains in which the concentration of carbon atoms becomes higher than average as a result of this fluctuation, there are also areas in which the carbon concentration is 6.7 per cent, i. e., up to the percentage of carbon in the cementite.

Thus, in the austenitic grains areas enriched in carbon up to the cementite concentration are constantly being formed on the one hand, and depleted of it so as to form a pure iron on the other hand.

On cooling the steel below the A_1 point, i. e., on supercooling it, some areas become centres or nuclei for the crystallisation of the ferrite-cementite conglomerate. What is the mechanism of this transformation?

Let us suppose that in some section of the austenitic grain there has been formed an area in which the carbon concentration is 6.7 per cent. In this area cementite molecules will form, i. e., the crystalline nucleus of cementite will arise. The adjoining areas of the austenitic grain will thus become depleted of carbon, with pure iron resulting. As the temperature is below 910° , in these areas there will occur a change of the space lattice of gamma iron into that of alpha iron, i. e., crystalline centres or nuclei of ferrite will be formed. The growth of each of the nuclei (cementite and ferrite) will stimulate the further growth of another nucleus; thus, the growth of the small crystal of cementite causes the further depletion of the adjacent areas of austenite in carbon atoms, thereby favouring the growth of the small crystals of ferrite. The growing ferrite crystals will reject the carbon atoms to their boundaries, thus providing the material for growth of the cementite crystals. This process, begun in some one place, will progressively occur in still greater volumes of the austenitic grain, resulting in the formation of a grain of the ferrite-cementite conglomerate.

The process of formation of the ferrite-cementite conglomerate grains usually begins simultaneously in several areas of an austenitic grain.

Such are the basic features of the mechanism of transformation of austenite into the ferrite-cementite conglomerate. One more question remains to be clarified. Fluctuations in the austenite composition take place at any temperature: why do these fluctuations not give rise to the formation of the ferrite-cementite aggregate at all temperatures, but only at temperatures below 723° ? This is explained by the fact that the free energy of austenite at temperatures above 723° (the A_1 point) is less than that of the ferrite-cement-

ite conglomerate, and consequently austenite is stable, whereas the ferrito-cementite aggregate is unstable. At the A_1 temperature the free energies of austenite and the ferrito-cementite conglomerate are equal, being thus in equilibrium; for this reason the transformation process of austenite into the ferrito-cementite aggregate cannot take place. Yet at temperatures below the A_1 , the free energy of ferrito-cementite conglomerate is less than that of the austenite, the former being therefore stabler, thus accounting for the unhindered change of austenite to the ferrito-cementite aggregate.

12. KINETICS OF THE TRANSFORMATION OF AUSTENITE INTO FERRITO-CEMENTITE CONGLOMERATE

The study of the kinetics of the austenite transformation into ferrito-cementite mixture, i. e., the determination of the temperature effect upon the rate of this transformation has been experimentally found to be most easily carried out not on continuous cooling but at a constant temperature, i. e., under *isothermal* conditions.

Experimentally this is accomplished in the following manner. Small specimens of the steel to be tested are heated to some temperature above the A_1 point, and after some period of holding at that temperature in order to be heated throughout they are rapidly transferred into a bath which has been heated previously to some temperature below the A_1 point. Consequently, the process of transformation of austenite into ferrito-cementite conglomerate will take place when the specimen is in this bath. As has been found for steel, the best method of recording the transformation of austenite into ferrito-cementite conglomerate is the magnetometric one. The *kinetic curves* representing the change of austenite to ferrito-cementite aggregate have been constructed by means of this magnetometric method.

Fig. 15 gives the curve showing the *isothermal transformation* of austenite into ferrito-cementite conglomerate in eutectoid steel at a given temperature. From this curve it may be seen that the transformation process does not begin as soon as the steel has attained the given temperature, i. e., not immediately after the specimen has been transferred into the furnace of the magnetometer. The transformation does not occur for some time, which is measured by the length Oa . The transformation begins at point a and reaches completion at point b after a lapse of time determined by the length ab . The steel has austenitic structure up to point a , and the transformation occurs between points a and b , i. e., austenite and ferrito-cementite conglomerate exist simultaneously in the steel

structure; after point b has been passed only the ferrito-cementite aggregate remains.

Let us construct such isothermal curves for different temperatures and then proceed as follows. Let us take temperature and time as coordinates (Fig. 16), draw a number of isotherms (horizontal lines) corresponding to temperatures, for which these isothermal curves have been constructed (Fig. 15), mark on these isotherms points a (beginning of the transformation) and b (end of the transformation), then connect all points a and b with even lines. A summary graph illustrating the isothermal transformation of austenite

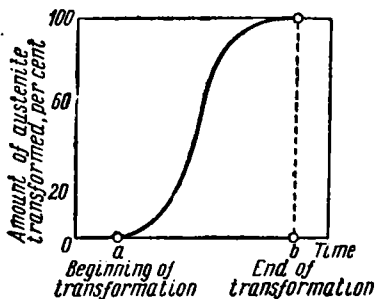


Fig. 15. Kinetic curve showing the isothermal transformation of supercooled austenite at a given temperature.

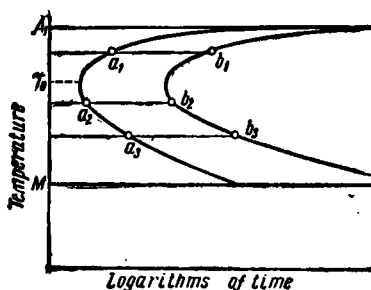


Fig. 16. Summary kinetic diagram showing the isothermal transformation of austenite.

will then be obtained. In order to limit the length of the diagram on the abscissa, time is not plotted on it, but a logarithm of time.

The left curve on this diagram determines the *beginning* of the transformation of austenite into ferrito-cementite conglomerate, while the right one shows the *end* of this change. The austenite exists in the region to the left of the left curve, the ferrito-cementite mixture in the field to the right of the right curve. The transformation process proceeds in the area located between these two curves. This means that in this area both austenite and the products of its decomposition, i. e., ferrito-cementite conglomerate exist simultaneously.

The line M represents the temperature of martensitic transformation. This line and the martensitic transformation itself will be discussed more amply in the following paragraphs.

A number of extremely important conclusions may be drawn from the diagram of the isothermal transformation of austenite (Fig. 16).

At temperatures below the bottom critical point A_1 , which corresponds to equilibrium, the supercooled austenite may exist for a length of time determined by the left transformation curve. The *degree of stability of the austenite* varies at different temperatures

of supercooling. The austenite is least stable when supercooled at temperatures within less than 100 to 150° of the equilibrium temperature A_1 . This minimum stability of the austenite is determined by the point T_0 of the inflexion of the left curve.

The austenite is found to be more stable at temperatures of supercooling higher than the T_0 point, as well as lower than this range. The stability of austenite is highest at temperatures but slightly below the A_1 point, on the one hand, and at those approaching the M temperature on the other hand.

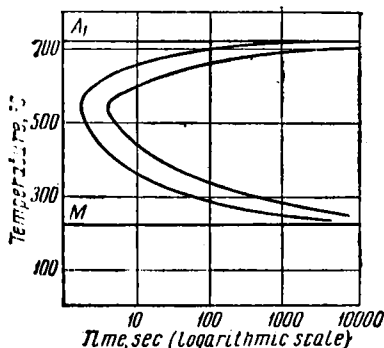


Fig. 17. Diagram showing the isothermal transformation of supercooled austenite in steel of grade Y8.

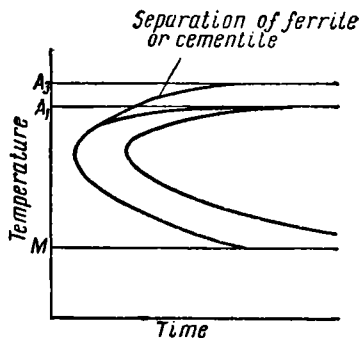


Fig. 18. Diagram showing the isothermal transformation of supercooled austenite in hypoeutectoid or hypereutectoid steels.

The real diagram of the isothermal transformation of supercooled austenite for the steel of grade Y8 is represented by Fig. 17. This diagram is similar to the schematic one which has just been considered. The diagram for the hypoeutectoid and hypereutectoid steels (Fig. 18) is similar, too, but with the addition of only one curve which represents the separation of an excess phase, i. e., of ferrite in hypoeutectoid steels or of cementite in hypereutectoid steels.

13. STRUCTURE AND PROPERTIES OF FERRITO-CEMENTITE CONGLOMERATE

The isothermal transformation of supercooled austenite at any temperature in the A_1 - M range gives rise to the formation of an aggregate composed of ferrite and cementite. But depending upon the transformation temperature, the structure and properties of this ferrite-cementite conglomerate are substantially different.

The ferrito-cementite aggregates formed at different temperatures of supercooling differ primarily in the degree of dispersion (refinement) of both phases which are constituents of this conglomerate, i. e., of ferrite and cementite. The lower the transformation temperature, the more dispersed are both phases. At high temperatures of transformation (at about 700°) the ferrito-cementite conglomerate becomes sufficiently distinct (differentiated) (Fig. 19). Such a well-defined aggregate composed of ferrite and cementite is called pearlite. At somewhat lower temperatures of transformation (at about

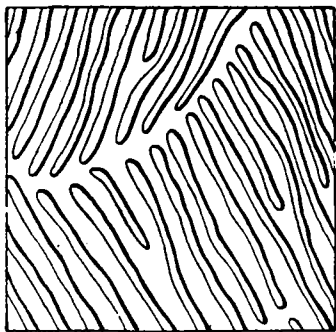


Fig. 19. Structure of lamellar pearlite at magnification of 2,500. Sorbite has the same structure at 5,000 $\times$, and troostite at 10,000 $\times$.

650°) the ferrito-cementite conglomerate is more dispersed. In this case it is called *sorbite*. At still lower temperatures of transformation (at 600° and down to the M range) the ferrito-cementite conglomerate becomes so dispersed that its structure can hardly be distinguished at any magnifications by an optical microscope. Only the photomicrographs made with an electronic microscope reveal that its structure is similar to that of pearlite and sorbite. Such highly dispersed ferrito-cementite aggregate is known as *troostite*.

The ferrito-cementite aggregates differ from each other not only in degree of dispersion of constituent phases but also in the form of the

cementite phase. The cementite phase in ferrito-cementite conglomerate may have either lamellar or globular (grained) form. The formation of the cementite phase of any kind depends upon the temperature to which the austenite has been heated prior to its transformation. It has been found that when austenite is heated up to comparatively high temperatures (exceeding the A_1 point by 100° or more) the ferrito-cementite conglomerate has a lamellar structure irrespective of the transformation temperature (Fig. 20, *a*). If the austenite has been heated to comparatively low temperatures (exceeding the A_1 point by not more than 50 to 60°) and the degree of supercooling has been relatively small, the cementite phase will have a grained structure (Fig. 20, *b*). Even at greater degrees of supercooling in this case the structure will become lamellar.

Ferrito-cementite aggregate of sufficiently distinctive (differentiated) structure with a cementite phase of grained form is called *grain pearlite* or, as some scientists prefer to call it, *grain cementite*. Ferrito-cementite conglomerates of more dispersed structure

with cementite in grain form have not been given a special name, and are called sorbite or troostite like the dispersed ferrite-cementite aggregates of lamellar structure.

The forming of a cementite phase of any kind in the ferrite-cementite conglomerate is explained by the degree of uniformity of the original (primary) austenite. The more uniform austenite (which corresponds to higher temperatures of heating it) is decomposed into an aggregate of lamellar structure. At lower temperatures of heating uniformity of structure is not attained; apart from the austenite, undissolved particles of cementite remain, which determine the grain form of the separated austenite.

The transformation of the supercooled austenite in the temperature range from A_1 to about 500° (Fig. 16) gives rise to the formation of polyhedrons or spherulites of ferrite-cementite conglomerate (Fig. 21), while the transformation in the temperature range from about 500° to M leads to the formation of a distinctive acicular structure (Fig. 22) similar to that of martensite. By studying the structure of these needles, in particular by means of an electronic microscope, it has been found that it is composed of the same aggregate of ferrite and cementite as is the troostite. Therefore this acicular structure is often called the *acicular troostite* (or sometimes *bainite*).

Besides the difference in microgeometry, there are also other differences between troostite and acicular troostite. First, the carbon content in the ferrite phase is somewhat different; the carbon content in the ferrite of troostite corresponds to that determined by the phase diagram, while the carbon content in the ferrite of acicular troostite is in excess of that for a condition of equilibrium. In other words, when the acicular troostite is formed not all the carbon contained in the austenite separates from the space lattice of iron. So the process of formation of the acicular troostite is partly one of diffusion and partly of non-diffusion.

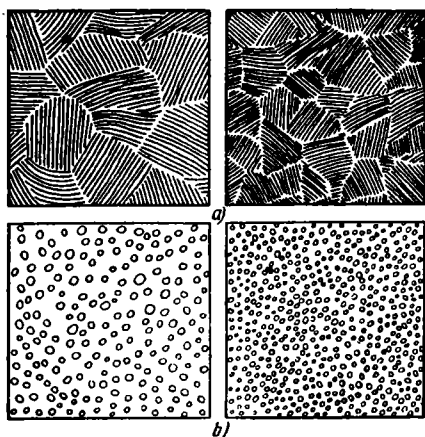


Fig. 20. Structure of the ferrite-cementite conglomerate in carbon steel in relation to the temperature of heating of the initial austenite and the transformation temperature.

a—lamellar structure: heating up to 900° and transformation at temperatures 710° (left) and 670° (right); b—grain structure: heating up to 780° and transformation at temperatures 710° (left) and 670° (right), $500\times$.

The second difference between troostite and acicular troostite is that the austenite is transformed into troostite through the generation of grains and their subsequent growth, whereas the change of austenite to acicular troostite is accompanied by hardly any growth of needles.

These two peculiar features of the formation of acicular troostite make it more approximate to martensite. According to these differences in the formation mechanism and in the microgeometry of the



Fig. 21. Globules of troostite formed in the upper zone of the supercooled austenite decomposition, 500 $\times$.

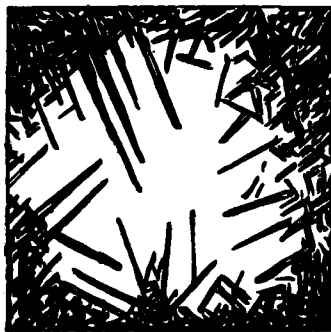


Fig. 22. Structure of acicular troostite, 500 $\times$.

products of austenite decomposition into ferrite and cementite, the whole temperature range from A_1 to M (designated as Ar) is divided into two sections: 1) the upper range (or upper step), from A_1 to about 500°, where pearlite, sorbite and troostite are formed; and 2) the lower range (or lower step), at about 500° to M , where the acicular troostite is formed. Transformation in the upper step is designated as Ar' and in the lower step— Ar'' . The properties of ferrite-cementite aggregates forming at different temperatures are different. The more dispersed the ferrite-cementite conglomerate, the higher are its strength, elasticity and hardness. As to plasticity and ductility, they improve first as the degree of dispersion increases, and begin to deteriorate after the maxima have been reached. Sorbite has the best values for plasticity and ductility (under otherwise equal conditions), these being somewhat higher than those of pearlite and considerably higher than those of troostite.

The Rockwell "C" hardness of eutectoid steel may be related to the degree of dispersion of the ferrite-cementite conglomerate by the following figures:

pearlite	10-20
sorbite	20-30
troostite	30-45
acicular troostite	45-55

14. THE MARTENSITIC TRANSFORMATION

When steel is cooled to the M temperature (see Fig. 16) or to still lower temperatures, the austenite changes not into the ferrito-cementite conglomerate but into martensite.

The transformation of austenite into ferrito-cementite aggregate consists, as has already been amply explained, of the following two simultaneous processes: 1) change of the space lattice of gamma iron to the space lattice of alpha iron; 2) separation of the majority of the carbon atoms from the space lattice of gamma iron and their precipitation into an isolated cementite phase. The second process consisting in migration of the carbon atoms inside the space lattice is thus a diffusion process. The first process, the change of the space lattice, is a non-diffusion process.

With high degrees of supercooling to comparatively low temperatures (200 to 300°) the space lattice changes freely. The second process, i. e., the separation of most of the carbon atoms from the space lattice of iron and their precipitation in the form of the cementite molecules, does not take place at low temperatures of supercooling for very obvious reasons that at comparatively low temperatures such as these the rate of carbon diffusion is very slow. The carbon atoms remain, therefore, in the space lattice of iron.

What is obtained as a final result? The fact that the carbon atoms remain in the space lattice of iron means that the martensitic structure obtained is a solid solution. But since martensite has the space lattice of alpha iron, it may be deduced that martensite is a solid solution of carbon in alpha iron.

The nature of martensite was completely unexplained not long ago. From the easily established fact of the martensitic structure of hardened steel being magnetic, the conclusion may justly be drawn that the space lattice of martensite is similar to that of alpha iron. On the other hand, it was experimentally proved long ago that the electrical resistivity of martensite is considerably higher than that of the ferrito-cementite conglomerate of any degree of dispersion. High values of electrical resistivity are characteristic of solid solutions. It therefore seems apparent that martensite is a solid solution of carbon in alpha iron. But this conclusion does not conform with the well known fact that alpha iron is hardly capable of dissolving the carbon. So contradictions arise. Many theories have been ad-

vanced to explain the nature of martensite. In books on metallography published in the twenties and even thirties of the current century a description of these theories may be found.

The Soviet scientists N. Y. Selyakov, G. V. Kurdjumov and N. T. Gudtsov were the first to discover the nature of martensite in 1927; they examined the structure of martensite by means of X-

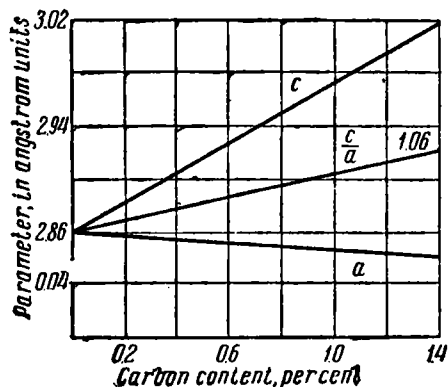


Fig. 23. Relation of parameters of the martensite space lattice to the carbon content.

rays. A. P. Gulyaev rightly claims that the works of these scientists may be considered as important as those of P. P. Anosov and were epoch-making in the field of steel metallography.

In their work, corroborated by many later investigations, it has been established that: 1) martensite is a single-phase system; 2) it has a tetragonal space lattice; and 3) the parameters of the space lattice are a linear function of the carbon content in it (Fig. 23). From this it has been deduced that the martensite does not

possess a stoichiometrical composition, i. e., it is not to be considered as a chemical compound. Yet what else can a single-phase system of varying chemical composition be, if not a solid solution?

The mechanism of martensitic transformation may be represented as follows. On rapid cooling, high internal stresses exceeding the low yield point of the austenite develop in the austenitic grains. This results in plastic deformation of the austenitic grains, i. e., some planes of the space lattice move in relation to each other, bringing about a slip. The space lattice becomes distorted and partly destroyed, and to replace it the iron atoms form another lattice which is more stable at temperatures below the A_1 point, i. e., the space lattice of alpha iron.

Thus, the elongated and narrow area of the austenitic grain, located along the line of slip with the space lattice of alpha iron, is actually a plate of martensite, which appears in the microsection as a long and thin needle. The acicular form is a very characteristic microstructural feature of martensite (Fig. 24).

The first needles of austenite pierce the entire austenitic grain almost throughout (Fig. 25, a). The rate of formation of martensitic needles is enormously high. The high velocity in taking photomicrographs (800 shots per second) of martensitic transformation,



Fig. 24. Structure of martensite in hardened steel of grade Y8, 500 X.

made by A. P. Gulyaev, did not show any growth of martensitic needles; a needle appears at once in some of the stills, though there is no sign of it on the preceding one, but the time interval between the two shots is only about 0.001 second. It may therefore be accepted that the martensitic needles appear practically instantly. The martensitic needles formed do not grow in length or breadth but remain invariable. As the martensitic transformation proceeds further new needles form (Fig. 25, *b*, etc.), all subsequent needles becoming shorter and shorter.

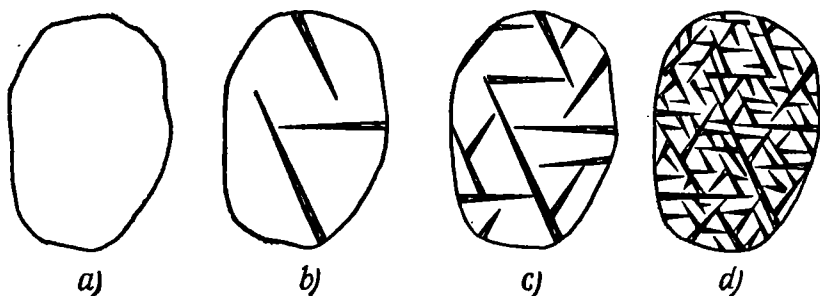


Fig. 25. Scheme showing the formation of martensitic structure. The figure represents one austenitic grain at several progressively lowering temperatures (from left to right) below the M point. After the entire volume of the grain has been filled with martensitic needles, the areas of residual austenite (light coloured) remain between them.

From the above discussed mechanism of martensitic transformation it follows that this process is non-diffusion, the chemical composition of austenite and that of martensite forming from it being completely identical.

Unlike the ferrite-cementite transformation, the martensitic transformation can hardly proceed isothermally. When steel is held at a constant temperature below the M point, the martensitic transformation ceases very rapidly. The resulting structure is made up of austenitic grains pierced by some quantity of martensitic needles. The temperature of the steel should be lowered progressively so as to enable the process to develop further. As the temperature is lowered, new stresses are set up in the austenitic grains, leading to the formation of new martensitic needles.

Thus, the martensitic transformation is accomplished through a temperature range. The temperature ranges $M-M_s$ (corresponding to the beginning and end of the martensitic transformation respectively) are determined by the carbon content in the steel (Fig. 26), i. e., the higher the proportion of carbon in the steel, the

lower the temperatures of the beginning and end of the martensitic transformation.

From the diagram in Fig. 26 it is clear that the martensitic transformation in steels with a carbon content exceeding 0.5 per cent is completed at temperatures below room temperature. This fact enables one to draw two very important conclusions. First, the structure of any hardened steel containing carbon in excess of 0.5 per cent is made up, if it is not cooled below room temperature, not only of martensite but also of some amount of austenite which has not been transformed. The higher the carbon content in the steel, the greater the amount of austenite will be. This austenite is called *residual*. Therefore when the structure of a steel containing carbon in excess of 0.5 per cent is said to be composed of martensite, the "residual austenite" is also implied. And secondly, in order to increase the amount of martensite in the structure of hardened steels with carbon in excess of 0.5 per cent and to increase thereby their hardness, it is sufficient to cool them below room temperature (to sub-zero temperatures). This is the main point of treating steel by cold, which will be dealt with in greater detail in Chapter VIII.

Some of the residual austenite remains in the structure of hardened steel even when this is cooled below the M_e temperature at which the martensitic transformation ends. This is because when austenitic grains are cooled to still lower temperatures they are broken by the martensitic plates into separate "sections", inside which the austenite is compressed from all sides. The specific volume of martensite being greater than that of austenite, this fact accounts for the hampering of the martensitic transformation by compression of the austenite in these "sections" on all sides.

Martensite may be obtained in the form of either coarse or fine needles. The degree of the acicularity of the martensite depends upon the size of the original austenitic grains: the larger the austenitic grains, the larger will be the needles of martensite. Since the austenitic grain size depends upon the excess of heating temperature above the A_1 point, the size of the martensitic needles may be used as a basis for determining the degree of steel overheating when

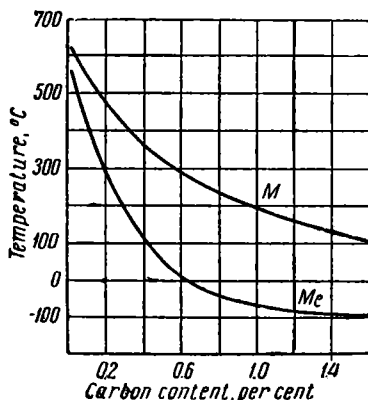


Fig. 26. Diagram showing relation of temperatures of beginning and completion of martensitic transformation in carbon steel to the carbon content.

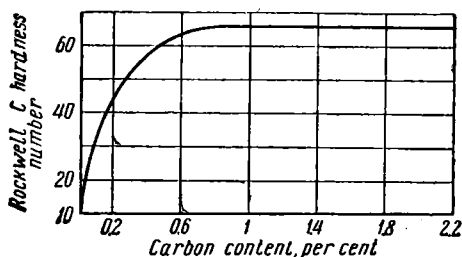


Fig. 27. Diagram showing relation of the hardness of iron-carbon alloys quenched so as to obtain the martensitic structure to the carbon content in these alloys (diagram as given by I. S. Gayev according to data of various investigators).

it is heated for quenching. In the structure of normally hardened hypereutectoid tool steel the martensitic needles cannot usually be detected because they are extremely fine.

The characteristic properties of martensite are its very high hardness (Fig. 27) and extremely low impact strength. The explanation of both these properties of martensite is the same, i. e., the considerable distortions of the space lattice of mar-

tensite, which hinder plastic deformation. The higher the carbon content in martensite, the greater the distortions of the space lattice, and, hence, the greater will be the resulting hardness. This conclusion is corroborated by Fig. 27.

15. TRANSFORMATION DURING CONTINUOUS COOLING

The study of the kinetics of the transformation of austenite under isothermal conditions has considerably reduced the experimental difficulties, for it made possible the elimination of the influence of the cooling rate. Because of this many laws governing this transformation have been established, the most important of them being already considered.

However, under real conditions of heat-treatment the transformations occur in most cases *not isothermally* but on *continuous cooling*. Therefore the task is to employ the diagram of isothermal transformation of supercooled austenite for the study of the transformations that take place under continuous cooling. This problem has been resolved by plotting the curves for continuous cooling on the diagram of the isothermal transformation of supercooled austenite. The lines for continuous cooling plotted on the diagram with coordinates temperature-time in logarithmic scale are represented by curves, but in order to simplify the plotting these curves will be considered as straight lines.

Several specimens of eutectoid steel are heated to a temperature above the A_1 point. This temperature is marked off in the diagram of isothermal transformation of supercooled austenite as point t (Fig. 28). Then the steel specimens are cooled at various cooling rates. The cooling processes are represented by the inclined straight

lines, which are the more steep, the faster the cooling rate. The slowest cooling rate (for annealing) is represented by the straight line v_1 , the slightly more rapid cooling rate (for normalising) by the straight line v_2 , the still more rapid cooling rate (for oil quenching) by the straight lines v_3 and v_4 , and the most rapid cooling rate (for water quenching) by the straight lines v_5 and v_6 .

On very slow cooling the straight line v_1 crosses both transformation curves as follows: the curve of beginning of transformation is

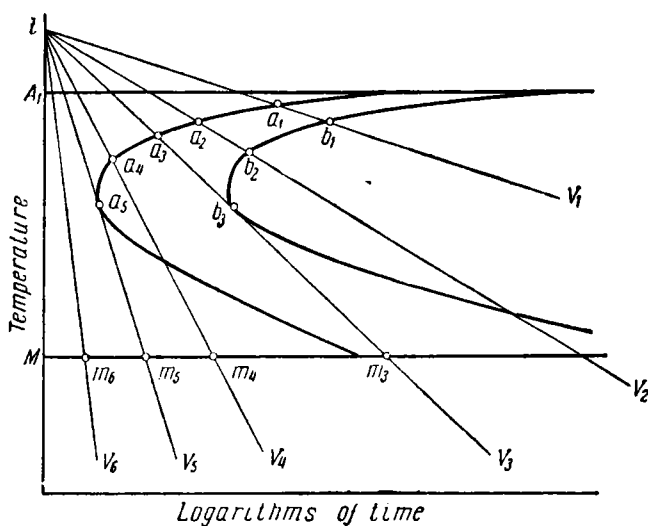


Fig. 28. Plotting of curves, illustrating continuous cooling, on the diagram of isothermal transformation of supercooled austenite.

intersected at point a_1 , and the curve of completion of transformation at point b_1 . This means that on slow cooling the austenite completely transforms into a ferrite-cementite aggregate. Since the transformation occurs at the highest temperatures (for the temperature range A_1 - M), the ferrite-cementite conglomerate becomes the least dispersed and sufficiently distinctive. This is pearlite.

On more rapid cooling (as in normalising) the straight line v_2 also intersects both transformation curves. This means that even in this case the austenite completely decomposes into ferrite-cementite conglomerate; but as the transformation range a_2 - b_2 corresponds to lower temperatures, the resulting aggregate is more dispersed. This is sorbite.

At any cooling rate not exceeding v_2 , the curves illustrating the cooling process intersect both transformation curves, which implies

the complete decomposition of the austenite into ferrite-cementite aggregate.

At a cooling rate faster than v_1 , e. g., at a cooling rate v_4 , the straight line v_4 , illustrating cooling, intersects only the curve corresponding to the beginning of transformation (the a_4 point), and does not cross the curve representing the end of transformation. This means that the ferrite-cementite transformation begins but does not reach completion. In other words part of the volume of the austenitic grains undergoes decomposition into ferrite and cementite, while for another part of this volume there is insufficient time for it to be transformed into a ferrite-cementite aggregate. This remaining part of the austenite, which has not been decomposed, undergoes the transformation into martensite on reaching the M temperature (the m_4 point). Thus, the structure of steel cooled at the rate v_4 consists partly of troostite and partly of martensite. Such a structure is peculiar to all steels cooled at a rate faster than v_3 , but slower than v_5 . For a carbon steel this cooling rate is equivalent to quenching in oil.

At any cooling rate faster than v_5 the transformation of austenite into ferrite and cementite does not take place at all, and the austenite completely transforms into martensite (the points m_5 , m_6). For carbon steel this cooling rate corresponds to the quenching in water.

In analysing the transformations taking place during continuous cooling one common failure, which is usually committed, must be avoided, when the diagram is considered in a formal way. This concerns the straight lines v_5 or others which are more sloping (between v_5 and v_1). These straight lines also intersect the M line showing the martensitic transformation; e. g., the v_5 line crosses the M line in the m_5 point. But from this it cannot be deduced that at such cooling rates the martensitic transformation does take place in the steel. In fact at the cooling rates equal to or less than v_5 the straight lines representing the cooling intersect both curves, and hence in steels thus cooled the ferrite-cementite transformation reaches completion. No austenite not already transformed remains, and consequently there is *nothing* to be transformed to martensite at the M temperature (the m_5 point). So under conditions of continuous cooling the m_5 point, and others located in the straight line M to the right of the m_5 point, have no physical sense.

The cooling rate v_5 , the least at which the martensitic areas appear in the steel structure, is called the *low critical quenching rate*. It is evident that the minimum critical quenching rate corresponds to the straightline illustrating the cooling, which is a tangent of the curve indicating the end of transformation. The cooling rate v_6 , being the least at which the ferrite-cementite conglomerate does not remain in the steel structure, which is composed only of martens-

ite (and unavoidable residual austenite), is called the *upper critical quenching rate*. Again it is clear that the upper critical quenching rate corresponds to the straight line for cooling, which is a tangent of the left curve indicating the beginning of the pearlitic transformation.

It should be noted here that by the critical cooling rate is usually understood precisely this upper critical quenching rate.

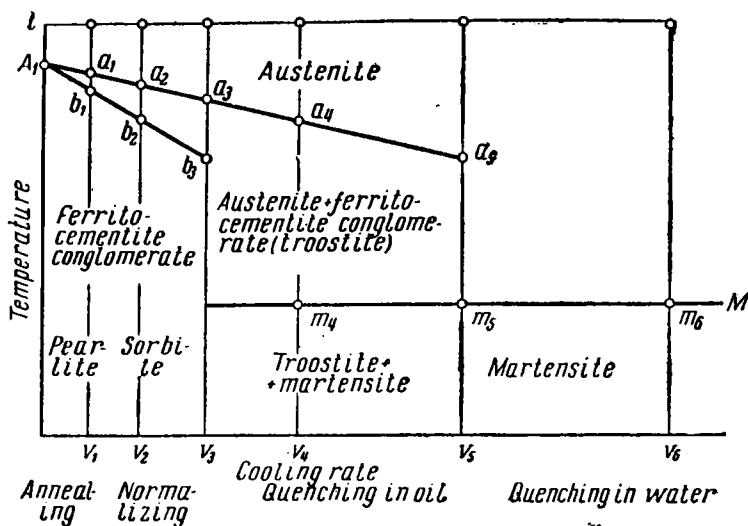


Fig. 29. Diagram showing relation of austenite transformation temperatures to the cooling rate (the letter symbols are the same as in Fig. 28).

The relation between the structures obtained on continuous cooling and the cooling rates can easily be made much clearer if the temperatures of beginning and completion of the martensitic transformation are plotted on the diagram with the temperature and cooling rate as coordinates (Fig. 29). From this information the diagram is easily understood.

16. PECULIARITIES OF THE TRANSFORMATIONS OF AUSTENITE IN ALLOY STEELS

In alloy steels the mechanism of the transformation of austenite into martensite and ferrite-cementite conglomerate does not qualitatively differ at all from that of the same transformations in carbon steels. The mechanism of the first (two-phase) transformation

differs only in that this transformation results in the formation not of an aggregate of ferrite and cementite but of a conglomerate made up of ferrite and carbides, because many alloying elements form complex carbides with cementite.

As to the kinetics of transformation, there are differences in this respect between carbon and alloy steels. For steels alloyed with elements that are not carbide forming (nickel, copper, aluminium) and containing small proportions of manganese (up to 2 to 3 per cent) the diagram for isothermal transformation of the supercooled

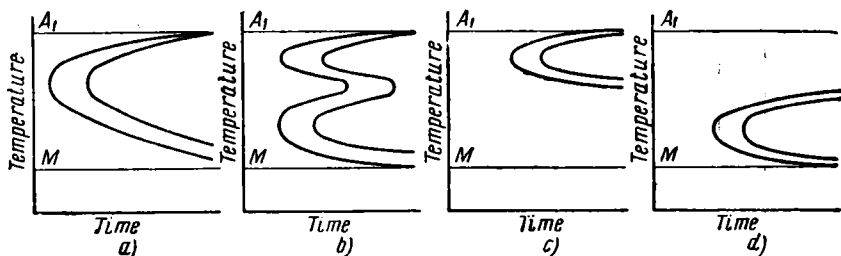


Fig. 30. Main types of diagrams showing the isothermal transformation of the supercooled austenite in alloy steels.

a—with only one step of transformation; b—with two steps of transformation; c—with only one upper step of transformation; d—with only one lower step of transformation.

austenite is of the same pattern (Fig. 30, a) as that for carbon steels, i. e., in the diagram there is only one minimum representing stable austenite.

For steels alloyed with carbide-forming elements (chromium, tungsten, molybdenum, vanadium) the diagram for isothermal transformation of the supercooled austenite possesses two minima indicative of the stable austenite (Fig. 30, b). The first (upper) minimum corresponds to the Ar' -transformation, i. e., the transformation of austenite into ferrite-carbide aggregate. The lower minimum corresponds to the Ar'' -transformation, i. e., the transformation of austenite into acicular troostite.

All alloying elements, both the carbide forming and those which do not form carbides (except cobalt) move the transformation curves to the right, this effect being the more pronounced the higher the content of these elements in the steel (Fig. 31). In other words, the alloying elements increase the stability of austenite.

Most alloying elements (chromium, tungsten, molybdenum, vanadium, silicon, aluminium) raise the temperature corresponding to the minimum stability of austenite, while some of them (nickel, copper, manganese) depress it (Fig. 31).

And finally, all alloying elements (with the exception of aluminium and cobalt) depress the temperature of the martensitic trans-

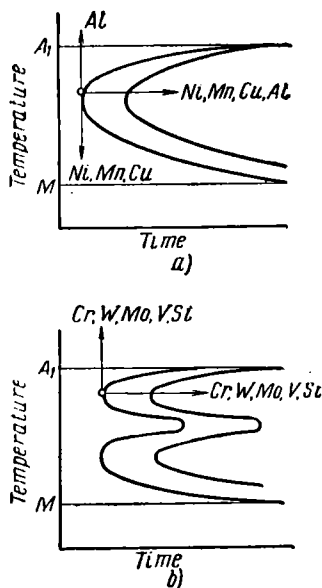


Fig. 31. Effect of alloying elements upon the diagram of isothermal transformation of supercooled austenite.

a—in steels with only one step of transformation; b—in steels with two steps of transformation.

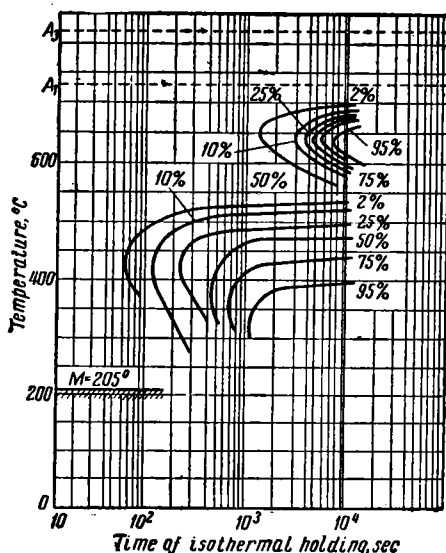


Fig. 32. Diagram showing the isothermal transformation of austenite in steel of grade 5XHB (A. A. Popov, V. S. Sagradze, S. E. Khorzov and E. F. Vostrikova).

formation, this effect being the more pronounced the higher the content of these elements in the steel. With a considerable content of alloying elements in the steel the temperature of the martensitic transformation may be below zero. This means that in such steels stable austenitic structure is obtained at room temperature.

The diagrams for the isothermal transformation of austenite in some complexly alloyed steels are yet more complex and unusual (Fig. 32). In particular in the diagram of the isothermal transformation of austenite for steels of type 18XHBA there is no upper minimum indicative of the austenite stability; for this reason the very long periods of holding are necessary for the Ar' -transformation to occur. This means, of course, that in such cases only the Ar'' -transformation (the formation of the acicular troostite) and the martensitic transformation take place at ordinary cooling rates.

These differences in the kinetics of transformation of austenite in alloy steels cause a number of peculiar features in the heat-treatment of such steels.

Let us compare the diagrams for the isothermal transformation of austenite for three steels of pearlitic class (Fig. 33): plain carbon steel (Fig. 33, *a*), low-alloy steel (Fig. 33, *b*) and high-alloy steel (Fig. 33, *c*). These diagrams differ from each other in that the curves of the ferrite-carbide transformation are moved to the right to a greater degree for the highly alloyed steels than for low-alloyed

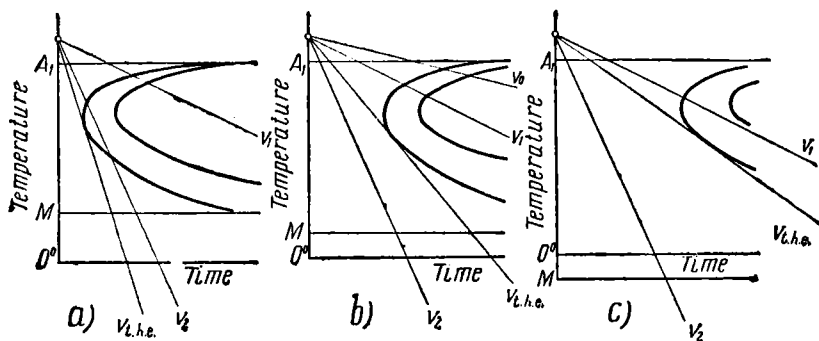


Fig. 33. Comparison between the diagrams of isothermal transformation of austenite for three differently alloyed steels.

ones (supposing, for the sake of simplification, that in these three steels there is no Ar'' -transformation).

Let us draw the straight lines showing the cooling in annealing v_1 on all these three diagrams. As is sufficiently well known, in the case of carbon steels the straight line for annealing intersects both transformation curves at the very top. This means that a clearly distinctive pearlite with a moderate hardness will be obtained in annealing the carbon steel. Since in many cases the specific purpose of annealing is to reduce the hardness of steel, this aim will be thus obtained.

The annealing of low-alloy steels presents quite another picture. The straight line for annealing v_1 (Fig. 33, *b*) intersects both transformation curves, but does so at lower temperatures. This means that as a result of the usual annealing of low-alloy steels a ferrite-carbide aggregate of a more dispersed structure (sorbite or even troostite) will be obtained, and therefore the hardness of annealed low-alloy steel will be markedly higher than that of annealed carbon steel; in most cases this is absolutely inadmissible and the annealing must therefore be accomplished at very low cooling rates (5 to 15 degrees per hour). It is impossible to accomplish such retarded annealing in a switched-off furnace which is allowed to cool down freely, for any heat-treating furnace cools at considerably

higher rates. Therefore the retarded annealing is usually done in a furnace which, though it is cooling, has at the same time been warmed up to a certain extent. The straight line indicative of this annealing process is shown in Fig. 33, *b*, being marked by the symbol v_0 .

In annealing highly alloyed steels the straight line characteristic of the usual annealing process at best intersects only one left curve, which indicates the beginning of the transformation (Fig. 33, *b*). This means that in the structure of annealed highly alloyed steels martensite will be found.

To cause pearlitic transformation in highly alloyed steels such low cooling rates that cannot be obtained in practice are necessary. How, in such cases, can the reduction of hardness of a highly alloyed steel be achieved? It may be achieved not by annealing but by the high tempering, in which the spheroidising and coalescence of carbides occur, i. e., the reduction of the degree of dispersion.

The degree of stability of austenite increases with the concentration of alloying elements; this fact, being the cause of many difficulties experienced in annealing, proves to be very favourable for hardening.

On the same three diagrams there are three straight lines v_2 corresponding to quenching in oil. The straight line v_2 for carbon steel (Fig. 33, *a*) intersects only the left curve indicating the beginning of transformation. This means that both troostite and martensite will be found in the structure of a carbon steel quenched in oil. The same straight line v_2 on the diagram for the low-alloy steel (Fig. 33, *b*) passes to the left of the left transformation curve. This means that an alloyed steel quenched in oil has a martensitic structure. Yet quenching in oil has a tremendous technological advantage over water quenching, since the internal stresses and strains usually set up by the rapid cooling are considerably reduced thereby; consequently, the danger of warping and formation of quenching or water cracks is materially reduced.

Highly alloyed steels quenched in oil (some of them even after normalising) possess an austenitic structure. This is explained by the fact that the M point of martensitic transformation lies below room temperature (Fig. 33, *c*).

17. THE REAL SPEED OF COOLING AND THE CRITICAL COOLING RATE OF QUENCHING

Cooling is the most important part of any heat-treating operation. The cooling rate determines the kind of structure resulting, and thus the specific properties of the steel obtained. The cooling rate is especially important in hardening. In order to obtain a mar-

tensitic structure on hardening the real speed of cooling must be higher than the critical quenching rate.

If the numerous but secondary factors are neglected the *real cooling speed* mainly depends upon the *weight of article* to be hardened, its *surface area* and the *nature and temperature* of the quenching or *cooling media*. Therefore it is wrong to say that the faster speed of cooling is always obtained on quenching in water, while the lower on quenching in oil. The speed of cooling of a massive article in water may be lower than the critical cooling rate, and martensitic structure will not be obtained. On the other hand, the speed

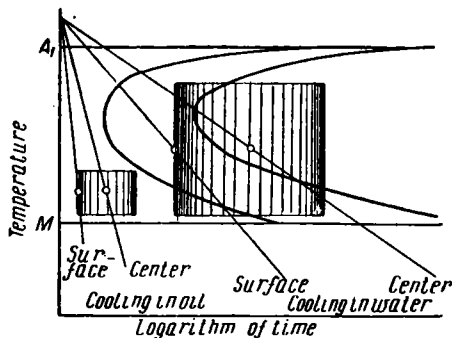


Fig. 34. Effect of a weight (mass) of articles made from the same steel upon the type of transformation of austenite; the smaller article is cooled in oil, the larger one in water.

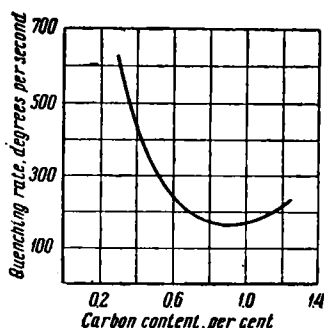


Fig. 35. Diagram showing the relation of the critical cooling rate of quenching carbon steels to the carbon content in them (After S. S. Steinberg).

of cooling of a small article in oil may be higher than the critical cooling rate in quenching, not only on the surface but also in the central portion, so that the article will be hardened throughout and martensitic structure will result (Fig. 34).

When quenching in water is said to be necessary for obtaining martensitic structure this refers only to relatively small articles and tools, the weight of which does not exceed a few kilogrammes. It is usually precisely such articles and tools that have to be quenched, and for them the real speed of cooling in water is always higher than the critical cooling rate of quenching.

The *critical cooling rate of quenching* also depends upon a number of factors. The chemical composition of steel is the main factor determining the value of the critical cooling rate of quenching. The relation of the critical cooling rate of quenching carbon steels to the carbon content is shown in Fig. 35.

For some alloy steels the values of critical cooling rate (according to Esser) are shown below:

Carbon content, per cent	Alloying element, and its proportion	Upper critical cooling rate in quenching, degrees per second
0.42	1.12 Ni	450
0.40	4.8 Ni	85
1.09	0.53 Cr	300
1.02	1.04 Cr	100
0.92	2.00 Cr	25
1.19	3.03 Cr	20

The critical cooling rate of quenching depends to a considerable degree upon the *hardening temperature*; the higher the heating temperature for hardening, the more uniform becomes the austenite, the larger its grains, the stabler it will be and, consequently, the lower its critical cooling rate of quenching. The critical cooling rate of quenching depends to a certain extent upon the *metallurgical nature of the steel*; the cleaner the steel with regard to various non-metallic inclusions, the less, consequently, the number of centres or nuclei, in the steel structure facilitating the *Ar*-transformation, the stabler the supercooled austenite in the upper step, and thus the slower the critical cooling rate of quenching.

The critical cooling rate in quenching does not depend, however, upon the weight of articles. Unlike the critical cooling rate, *the critical speed of quenching is an intrinsic property of the steel itself.*

18. HARDENABILITY

When articles are quenched the cooling rate decreases from the surface to the interior, being the highest at the surface and the lowest in the central portion.

If the cooling rate in the central portion of an article is in excess of the critical one for the given grade of steel, the article will be hardened through its entire cross-section, i. e., hardened throughout. But if the cooling rate in the centre of the section is lower than the critical one, the central portion remains unhardened.

The hardenability is of great importance for obtaining high and, what is more important, uniform properties across the whole cross-section of an article.

Suppose that three equal-size articles made of three different grades of steel possessing different hardenability have been hardened (Fig. 36), and that two of them (*a* and *b*) have not been hardened throughout (the hardened zone is shaded), whereas the third (*c*) has. The hardened zones will have the martensitic structure, while the structure of non-hardened zones will be troostite or even sorbite.

unit (Fig. 39), where it is immediately quenched by water spraying.

Two opposite sides of the cylindrical surface of the specimen are then ground so as to obtain two longitudinal areas sunk by 0.2 to 0.5 mm, and at these surfaces the Rockwell "C" hardness is determined, the measurements beginning at a distance of 48 mm from the hardened end. The first ten measurements in the direction of the

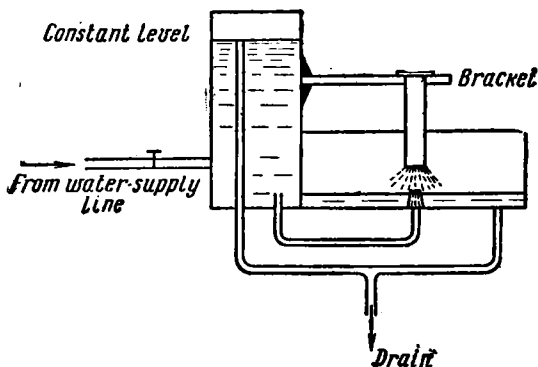


Fig. 39. Sketch of the unit for determining hardenability of steel with the end-quench test.

hardened end are made every 3 mm, and the remaining 13 measurements every 1.5 mm. For each two measurements, taken at opposite sides, the arithmetical mean is calculated.

The hardenability test results for the steel are represented in a graphical form (Fig. 40). From the hardenability test results obtained by the end-quenching method, the diameter of the cylindrical part having the same hardness in the centre when quenched either in water of temperature 10 to 20° or in oil with temperature of 15 to 70° may be determined, the evaluation being made on the basis of the diagram shown in Fig. 41.

The evaluation of the hardenability of carbon tool steels is made in accordance with the so-called fracture scale (Fig. 42) specified by the U.S.S.R. State Standard. The hardenability of the steel is estimated according to the fractures of three specimens having a cross-section of 20 by 20 mm, or round specimens 21 to 23 mm in diameter and 100 mm long, notched at the centre (the depth of a notch being 2 to 5 mm, or 5 to 7 mm for key-hole type). The specimens are first annealed at a temperature range of about 730 to 750° and then, after being held for 2 hours at this temperature, they are cooled with a furnace down to 650° C. They are then quenched in water of 10 to 30° from temperatures of 760, 800, and 840° C.

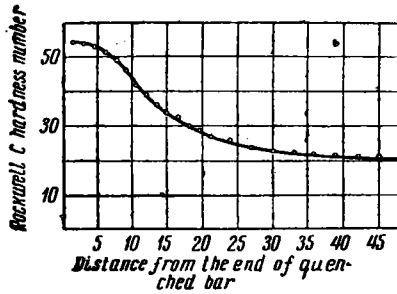


Fig. 40. The hardenability curve.

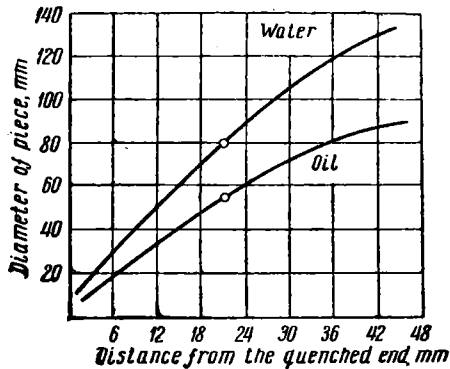


Fig. 41. Diagram illustrating the graphical method for determining, on the basis of the hardenability curve, the diameter of a cylindrical article having a given hardness after quenching.

Example: From the data obtained in the end-quench test it has been established that the hardness $40R_C$ is located at a distance of 21 mm from the quenched end. The diameter of a piece having the same hardness in the centre is 80 mm when quenched in water, and 55 mm when quenched in oil.

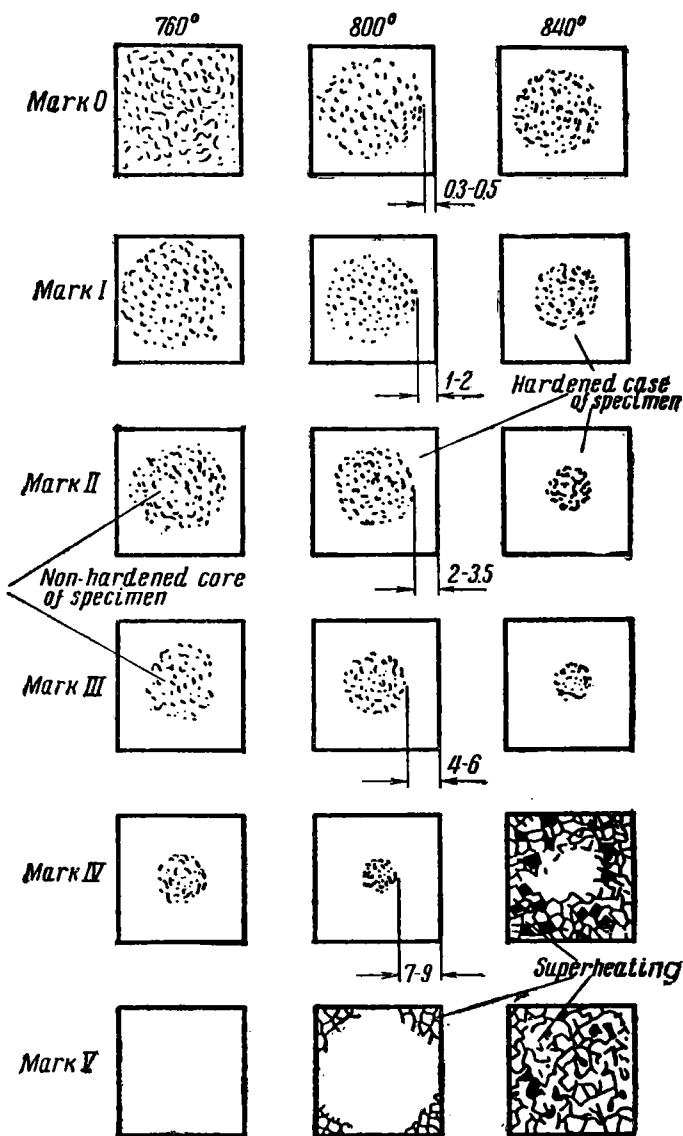


Fig. 42. The hardenability scale for carbon tool steels (according to State Standard).

19. COOLING MEDIA

Various liquid coolants are used in quenching steels as the principal quenching (or cooling) media.

Let us consider the process of cooling by immersion in a cooling liquid an article which has been heated to a high temperature. When it is immersed in a liquid coolant the liquid layer immediately surrounding it will almost instantly be heated up to boiling point and vaporised. The article will be enveloped in a steam layer or vapour coating, which will prevent further access of the cooling liquid to the surface of the article. The surface layers of the article will be cooled very intensively before the formation of the vapour coating.

To evaporate a liquid substance it is necessary:

1) to heat the liquid from its original temperature t_0 to boiling point t_{boil} ; the amount of heat required for this purpose is proportional to the specific heat of the liquid c_{liq} and temperature difference $t_{boil} - t_0$:

$$c_{liq} (t_{boil} - t_0);$$

2) to heat the liquid up to boiling point t_{boil} in order to change it into the gaseous state; the amount of heat required for this purpose is proportional to the latent heat of vaporisation of the liquid L_{vapour} ;

3) to heat the vapour from the boiling point of the liquid to a certain final temperature t_f ; the amount of heat required for this purpose is proportional to the specific heat of vapour c_v and temperature difference $t_f - t_{boil}$:

$$c_v (t_f - t_{boil}).$$

The vapour coating formed is continuously washed off by the surrounding layers of liquid, this effect being the more pronounced the less the viscosity of the liquid, i. e., the greater its fluidity.

When an article is cooled below the boiling point of the liquid the vaporisation process comes to an end, and the heat of the article is transferred to the liquid through thermal conductivity only, the further cooling being the more drastic, the higher the heat conductivity.

So the quenching power or withdrawal of heat of a cooling liquid depends upon: a) its initial temperature; 2) its boiling point; 3) the specific heat of the liquid; 4) the latent heat of vaporisation; 5) the specific heat of vapour; 6) the thermal conductivity; 7) viscosity.

The quenching media most widely used in commercial practice are water and mineral oils. The cooling efficiency of the mineral oils is by many times less than that of water because their specific

heat, latent heat of vaporisation and thermal conductivity are lower than those of water, while the viscosity is considerably higher, as is evident from Table 5 compiled by D. Y. Vishnyakov.

Table 5

Comparative specifications for physical properties of quenching media
(values for water are assumed to be 100 per cent)

Quenching liquid	Boiling point	Latent heat of vaporisation (total)	Thermal conductivity	Specific heat	Viscosity	Hardness at the surface of quenched steel
Water	100	100	100	100	100	100
Machine oil of grade I	330	45	17	46	440	49
Spindle oil of grade II	330	45	17	48	320	49

In order to obtain the martensitic structure in hardening carbon and most low-alloy steels, conditions should be created which enable the quenching to be carried out at a very high cooling rate (faster than the critical rate) through the range of *Ar*-transformation, where the stability of the austenite is at its lowest value (650 to 550°); this prevents the transformation of austenite into a ferrite-carbide conglomerate. On the other hand, it is desirable to cool an article at a considerably slower rate within the range of martensitic transformation (300 to 200°). In this case the internal thermal stresses, which invariably develop, are bound to be correspondingly lower, which is especially important for the steel in this temperature range since its ductility decreases drastically. A liquid that would permit the creation of such cooling conditions might be termed an ideal quenching medium. Those liquids which are in fact used commercially in quenching do not conform to at least one of the above-mentioned conditions for an ideal quenching medium, as is clear from the data of S. S. Steinberg, represented in Table 6.

From Table 6 the following main conclusions may be drawn:

1) when cooling in fresh water, a high cooling rate is obtained both within the range of the *Ar*-transformation (which is excellent) and within the range of the martensitic transformation (which is very bad);

2) on the other hand, on cooling in oil a slow cooling rate is obtained within the *Ar*-range, which is insufficient to suppress the trans-

Table 6

Cooling Rates for Small Specimens in Various Cooling Media

Cooling medium	Cooling rate (degrees Centi- grade per second) in the temperature range	
	650-550°C	300-200°C
Water at 18°	600	270
" " 30°	500	270
" " 50°	100	270
" " 75°	30	200
10 per cent NaOH water solution	1200	300
" " NaCl " "	1100	300
" " Na ₂ CO ₃ " "	800	270
" " H ₂ SO ₄ " "	750	300
Distilled water	250	200
Water saturated with CO ₂	100-200	200
Soap suds	30	200
Oil emulsion in water	70	200
Mineral oils (average)	100-150	20-50
Copper plates	60	30
Steel plates	35	15
Compressed air	30	10
Still air	3	1

formation of austenite in carbon steels into a ferrite-cementite aggregate, as well as within the martensitic transformation range, which is considered as a fairly favourable condition;

3) the warming of water markedly reduces the cooling rate within the *Ar*-range, and has very little effect on the cooling rate in the martensitic transformation range; from this it follows that in most cases the preheating of the water is not advisable; it should be noted though, that the warming of oil decreases its viscosity and somewhat increases the cooling rate within the *Ar*-range; hence oil is often warmed up to 40 to 80° prior to quenching;

4) matters dissolved in water (sodium hydroxide, sodium carbonate, sodium chloride and others) increase the cooling rate, i. e., are conducive to greater severity of quenching;

5) gases absorbed in water, on the other hand, markedly reduce the cooling rate; this explains why fresh water cools less drastically than already used water which has lost part of the gases dissolved in it;

6) scum-forming substances (e.g., soap) drastically reduce the cooling rate.

Apart from by the use of liquids, the cooling of a tool to be quenched may in some cases be done by directing jets of compressed air against it. Cooling by compressed air is used in particular in

quenching small tools made of high-speed steels. Cooling between steel and/or copper plates is employed in cases where warpage of thin articles in hardening must be avoided (e.g., gears, disc or circular saws, etc.).

Chapter IV

TRANSFORMATIONS IN HARDENED STEEL ON HEATING (THEORY OF TEMPERING)

20. CHANGES IN STRUCTURE OF HARDENED CARBON STEEL ON HEATING

The structure of carbon steel quenched from a temperature above the A_1 - or A_{cm} -ranges consists mainly of tetragonal martensite and a small amount of residual austenite. In the structure of the steel quenched from a temperature of between A_1 and A_2 , or A_1 and A_{cm} , there is also ferrite (in hypoeutectoid steels) or cementite (in hyper-eutectoid steels). The tempering operations dealt with in this chapter are not affected by the presence of ferrite or cementite in the steel structure; for this reason these structural constituents will not be taken into consideration.

Both the structural constituents of hardened steel, tetragonal martensite and austenite, are unstable. If this is so, the hardened steel should naturally have a potential tendency to pass into an equilibrium or stable condition. But this equilibrium state cannot be reached at room temperature because of the low mobility of atoms (slow rate of diffusion). As the temperature rises the mobility of atoms increases (the diffusion rate accelerates), and in the unstable structure of the hardened steel the phase and structural changes conducive to an equilibrium state may actually take place. This is the essential nature of the tempering process.

The transformations that take place in the structure of the hardened steel on tempering cannot be detected by means of an optical microscope. They have been examined mainly by two methods—dilatometric and X-ray tests. Recently scientists have also employed the newly developed electron microscopes in their investigations.

The dilatometric method consists of studying the linear or volumetric expansion or contraction of the hardened steel to be tempered as the temperature rises. Since martensite has the maximum specific volume in comparison to all other phases, and the austenite the minimum one, it is evident that the linear (or volumetric) contraction of a specimen of hardened steel is indicative of the decomposition of the martensite, whereas its linear (or volumetric) expansion illustrates the decomposition of the austenite.

An experimental study of the linear expansion or contraction of a specimen of hardened steel in tempering (Fig. 43) has shown that the specimen length first decreases (in the temperature range of 80 to 200°) and then increases (in the temperature range between 200 and 300°); finally, beginning at the temperature of 300°, it again decreases continuously. By comparing this dilatometric curve with data of the X-ray test it has been possible to ascertain that four transformations take place in hardened steel during tempering:

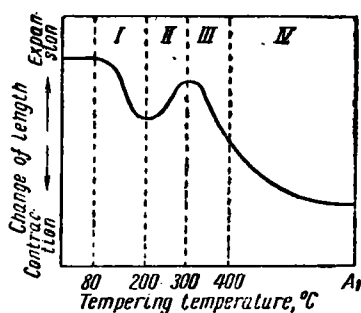


Fig. 43. Change in length of the hardened specimen in tempering. The temperature zones of four types of transformations in tempering are indicated with Roman numerals.

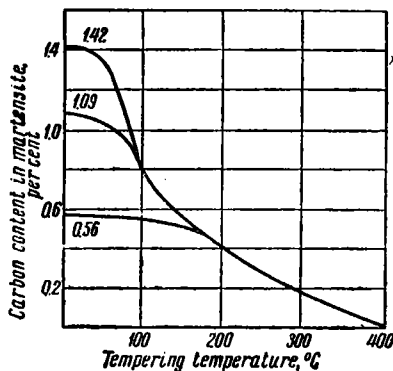


Fig. 44. Change in carbon content of martensite on tempering (according to E. Z. Kaminsky and T. I. Stelletskaia).

- 1) separation of carbon atoms from the space lattice of tetragonal martensite;
- 2) decomposition of the residual austenite;
- 3) transformation into cementite;
- 4) coagulation of cementite.

The first transformation, taking place mainly in the first temperature range (80 to 200°), consists in the formation of submicroscopic cementite particles (not visible under a microscope) by the carbon atoms being gradually thrown out from the space lattice of the tetragonal martensite. The martensite becomes depleted of carbon (Fig. 44). This depletion leads to the gradual decrease of its degree of tetragonal pattern ($\frac{c}{a}$) which approximates to 1. This martensite, being gradually depleted of carbon and having a space lattice which is also gradually changing from a tetragonal to a cubic pattern, is called an alpha solution. In so far as the specific volume of the alpha solution is less than that of tetragonal martensite, the linear contraction occurs in the first temperature range (Fig. 43).

The first transformation, which proceeds intensively in the first temperature range (80 to 200°), covers the second temperature range (200 to 300°), decreasing throughout the range and ceasing completely at a temperature of about 400°. At this temperature the alpha solution almost completely loses its carbon, and its space lattice assumes a cubic pattern. The alpha solution turns into ferrite. To the end of the first temperature range (200°) the structure of the steel to be tempered is composed of the alpha solution (considerably depleted of carbon but still retaining some degree of the tetragonal pattern), the submicroscopic cementite particles separated from it, and the residual austenite. Such heterogeneous structure is sometimes called *tempering martensite*.

In the *second temperature range* (200 to 300°) the first transformation is simultaneously accompanied by the second, i. e., the decomposition of austenite into an aggregate composed of alpha solution and cementite. The alpha solution obtained as a result of the breaking down of the austenite is similar to that which has been obtained at the given stage from the tetragonal martensite; it still holds some amount of carbon dissolved in it, and its space lattice continues to be tetragonal. As the specific volume of austenite is less than that of the products of its decomposition, the second transformation is accompanied by the expansion of the specimen (Fig. 43).

The second transformation reaches completion at the end of the second temperature range, and the structure of the steel tempered to 300° consists of alpha solution and cementite.

The *third transformation*, which takes place partly in the first and the second temperature ranges but proceeds especially vigorously in the *third temperature range* (300 to 400°), involves the cementite phase. The nature of the third transformation is not yet entirely clear. Some scientists believe that the chemical composition of the submicroscopic cementite particles that are thrown out of the tetragonal martensite and alpha solution in the first and second temperature ranges is different from that of the cementite Fe_3C . But the chemical composition of this cementite has not yet been determined exactly. Therefore its composition is designated by the formula Fe_xC . According to these scientists, the transformation of the cementite Fe_xC into the ordinary cementite Fe_3C takes place in the third temperature range.

Other scientists believe that at any tempering temperature the ordinary cementite Fe_3C separates from the tetragonal martensite and alpha solution, but the particles of the former do not separate completely from that alpha solution from which they have been thrown out; the boundary atomic layers in the space lattices of the alpha solution and cementite are common to both phases. The third

transformation, according to these investigators, consists in the isolating of the cementite particles.

The first transformation ends in the third temperature range, which brings about, as does the first temperature range, the linear contraction of a specimen. The third transformation (the cementite transformation) probably leads also to the same thing.

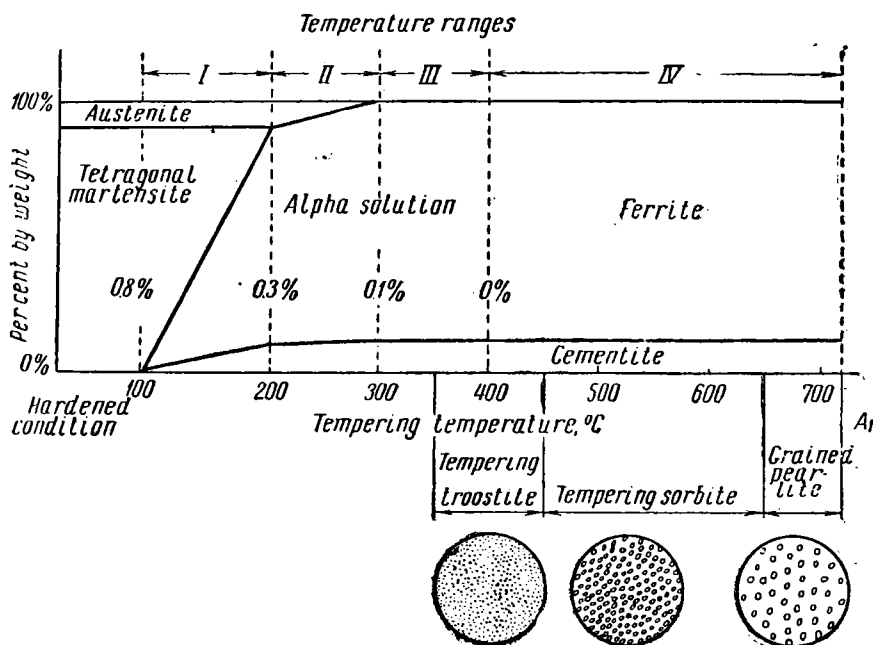


Fig. 45. Diagram showing the structural changes in hardened steel on tempering. The temperature ranges of the tempering process are indicated with Roman numerals. The figures indicate the carbon proportion in martensite. The structure patterns of troostite, sorbite and grained pearlite are shown in circles schematically.

In the *fourth temperature range* (from 400° to A_1) there occur a number of various transformations, which taken in combination are called the fourth transformation. In the fourth temperature range the processes of relieving the elastic strains and recrystallization of ferrite plastically deformed as a result of the prior volumetric changes take place. The further linear contraction of the specimen results from these processes, and simultaneously the coagulation of the cementite particles also takes place.

The coagulation or coalescence of the cementite particles proceeds continuously through the entire first, second and third temperature ranges. But in the *fourth temperature range* the process of cementite

both raise the temperature of ferrite recrystallisation and inhibit the process of carbide coagulation.

The combined effect of all this is that the hardness of steels alloyed with carbide-forming elements is higher than that of carbon steels tempered at the same temperatures (Fig. 47). Usually this may be briefly formulated as follows: the alloy steels, especially those alloyed with carbide-forming elements, are *less sensitive to tempering*. With some highly alloyed steels, for example high-speed steels, this

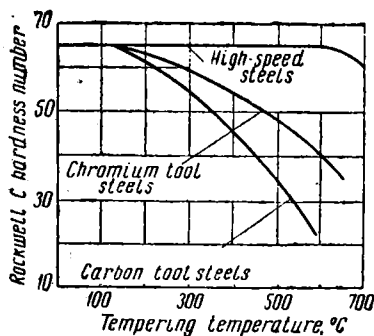


Fig. 47. Effect of a tempering temperature upon the hardness of hardened carbon and alloy tool steels.

property, i. e., the ability to retain hardness at high temperatures, is especially pronounced. The martensite in such steels undergoes the initial steps of decomposition corresponding to a maximum hardness only at 550 to 560° (it occurs in carbon steels at 150°).

An interesting feature of some highly alloyed steels is their ability to develop the so-called *secondary hardness*. This is manifested by the hardness of the steel being not only decreased upon tempering to some temperatures (between 500 to 550°), but even markedly increased as compared with the initial value.

This phenomenon is explained not only by separation of the carbide in a highly dispersed condition (as has been already pointed out), but also by the decomposition of the residual austenite, possessing a moderate hardness ($H_B=200$) as compared with the very high hardness of the tempering martensite ($H_B=650$ to 750).

The increase in hardness as a result of the austenite decomposition is common to all steels, but it is scarcely noticed because the amounts of residual austenite in hardened carbon and low-alloy steels are small. On the other hand, the amounts of residual austenite in highly alloyed steels are very significant (up to 30 per cent and over), the effect of hardness increase as a result of its decomposition being noticeable to a high degree.

23. TEMPER BRITTLINESS

Temper brittleness is an interesting feature peculiar to the tempering process applied for some hardened alloy steels.

As the tempering temperature of hardened carbon steels rises, their impact strength continuously increases (see Fig. 46). The relationship of impact value to the tempering temperature of certain hardened

alloy steels is expressed by a more complex curve (Fig. 48). This curve is remarkable because it has two minima (lowest points): the first at a temperature of about 300° and the second at about 550° . The first minimum value of impact strength noticed at the temperature of 300° is termed *temper brittleness of the first kind*, while the second minimum, at 550° , is *temper brittleness of the second kind*.

At the present time the causes of temper brittleness of the first kind are not clear. It is supposed that it is caused by the separation of carbides in lamellar form from the alpha solution in the relevant temperature range, which is detrimental to the impact resistance of the steel. At higher tempering temperatures the carbides are precipitated in grain form.

The nature of temper brittleness of the second kind is completely different. First, it is peculiar to by no means all steels, but only to a group of them, particularly chromium-nickel and chromium-manganese steels. Secondly, it is noticed only when a steel has been slowly cooled after tempering (Fig. 48). In rapid cooling after tempering there is no second minimum of the impact value. Here is an example of data gained by G. V. Uzhik and A. A. Zuikova, compiled for a steel of grade 30XPC that has been quenched in oil and subsequently tempered at a temperature of 640° .

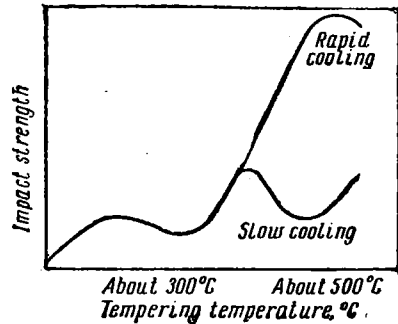


Fig. 48. Diagram showing the relation of the impact strength in hardened alloy steel to a tempering temperature.

Cooling after tempering	Tensile strength	Yield point	Elongation	Contraction	Impact value
in oil	93.0	78.0	20.4	57	14.5
with a furnace at . . .					
a rate of 20					
degrees per hour	88.0	71.5	20.7	56	3.7

It is interesting to note that if steel which has been slowly cooled after tempering and has a consequently low impact value is tempered a second time at the same temperature with subsequent rapid cooling, the impact value will be high. On the other hand, if steel which has been rapidly cooled from a tempering temperature and has a consequently high value of impact strength is tempered a second time at the same temperature with subsequent slow cooling, then its impact value will be low. Hence an *impact resistance of the second kind* is called a *reversible one*.

Temper brittleness of the second kind is also found when a steel subject to temper brittleness is held for a long period of time at a temperature of about 500° (Fig. 49) and then cooled at any rate. *The brittleness manifested by a steel as a result of its being held for a long time at high temperatures is termed hot brittleness.*

The causes of temper brittleness of the second kind have not yet been ascertained experimentally. The most plausible supposition is that on slow cooling from the tempering temperature of between 400

to 600° some brittle chemical compounds, probably lamellar carbides of the cementite type (Fe , Mo), C , separate at the grain boundaries of

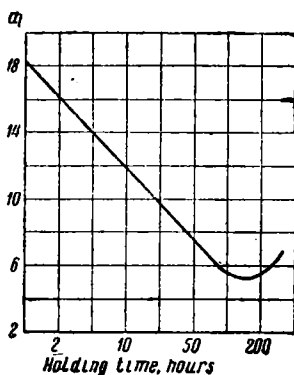


Fig. 49. Effect of the time of holding upon the impact strength of steel of grade 30XH, quenched from a temperature of 860° and tempered at 650° (according to V. I. Prosvirin and E. I. Kvashnin).

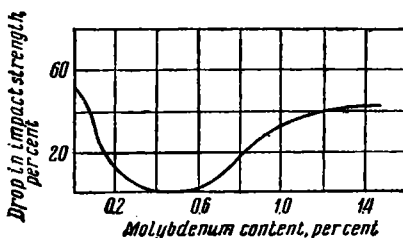


Fig. 50. Effect of molybdenum upon the decrease in impact strength of steel of grade 35XF after holding for 100 hours at a temperature of 500°C , as compared with the improved condition of the steel (according to V. I. Prosvirin and E. I. Kvashnin).

ferrite. On rapid cooling, however, there is insufficient time for these compounds to precipitate and they remain dissolved in the ferrite. The secondary increase in impact strength as a result of long holding at a temperature of 600° (see Fig. 48) is explained by the spheroidising of the lamellar products of precipitation.

Temper brittleness of the first kind is not essentially present in alloy structural steels, as these steels are in general tempered at considerably higher temperatures, usually at 550 to 650° . But this is by no means true of temper brittleness of the second kind, which manifests itself on cooling from precisely those temperatures at which structural steels are usually tempered. Therefore temper brittleness of the second kind is a very great inherent disadvantage of those alloy steels which are sensitive to it. When the sensitiveness of a steel to temper brittleness is spoken of, temper brittleness of the second kind is usually meant.

The most radical means of eliminating temper brittleness consists in alloying a steel with additional molybdenum (Fig. 50) or, less effectively, with tungsten. Chromium-nickel-molybdenum and chromium-manganese-molybdenum steels sharply differ from chromium-nickel and chromium-manganese steels in that they are completely free of the defect of temper brittleness.

The main groups of steels subject to temper brittleness are as follows:

- 1) manganese steels of higher manganese content (10Г2, 30Г2, and others);
- 2) chromium steels of higher carbon content (40X, 45X, and others), as well as IX15 and stainless steels;
- 3) chromium-silicon steels (35XC, 40XC, and others);
- 4) chromium-manganese steels (40X2Г, 15X2Г2Т, and others);
- 5) chromium-nickel steels (30XH, 40XH, 20XH3A, and others);
- 6) chromium-manganese-silicon steels (20XГC, 30XГC, and others);
- 7) manganese-silicon steels (25CГ, 35CГ, and others).

Steels which are not sensitive to temper brittleness, e. g.:

- 1) carbon steels;
- 2) manganese steels of moderate manganese content (20Г, 40Г, and others);
- 3) chromium steels of low carbon content (15X, 20X, etc. up to 35X inclusive);
- 4) molybdenum steels (15M, 30M and others);
- 5) chromium-molybdenum steels (30XMA, 35XM, and others);
- 6) chromium-molybdenum-aluminium steels (35XMIOA);
- 7) nickel steels (30H, 40H, and others);
- 8) chromium-nickel-tungsten steels (18XHB, 25XHB);
- 9) chromium-nickel-molybdenum steels (35XHM, 35XH3M, and others);
- 10) silicon steels (60C2) and others.

However, if the above-mentioned steels are kept for a long time at a tempering temperature and then slowly cooled they also will be sensitive to temper brittleness.

PART TWO

ELEMENTS OF THE PROCESS OF HEAT-TREATING

Chapter V

HEATING OF STEEL

24. FACTORS GOVERNING HEATING RATE

The heating of parts for heat-treatment can be accomplished in either of the following two principal ways:

1. Heating in a furnace previously heated to the heat-treating temperature (Fig. 51, *a*) or above it (Fig. 51, *b*).

2. Heating with furnace at required rate (Fig. 51, *c*).

In some cases a third method of heating (Fig. 51, *d*) is used, i. e., the primary heating is accomplished in a furnace heated to a temperature below that required for the heat-treating operation, the subsequent heating of parts is conducted together with that of the furnace after the first temperature has been reached.

The first method of heating is used in batch-type and continuous heating furnaces. The heating processes in the batch-type and continuous furnaces differ in that in the former there is a uniform temperature in the heating chamber, but in the latter the temperature varies with the length of the furnace, so that the article passes through the various zones heated to different temperatures as it moves through the furnace along its bottom.

The first method of heating (i. e., in a heated furnace) is employed more often than the second. It is preferred because the total time of heating is less. The advantages of rapid heating from the industrial point of view are evident: the productivity of the furnace is increased, the total time required for the heat-treating operation is reduced and the consumption of fuel or electric power is decreased. The advantages of a rapid heating are also unmistakable from the technological point of view: the less the period of time during which the steel is exposed to elevated and high temperatures, the less it is oxidised, the less is the extent of its surface decarburising, and the less will be the grain growth (when heated above the critical temperature).

The second heating method, i. e., deliberately slowed-down heating, is sometimes used because the heat transfer from the exterior to the interior of an article does not occur momentarily but at a certain definite rate, which is determined by the thermal conductivity of the steel. Therefore, the temperature of the central portion is always lower than that at the surface of an article (Fig. 51) during the

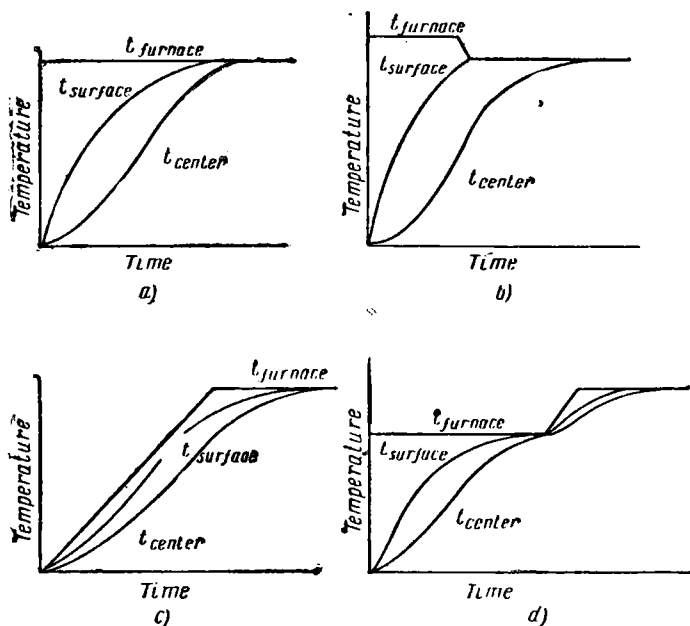


Fig. 51. Diagram showing the principal methods of heating.

a—heating in furnace at constant temperature, $t_{furnace}$; b—rapid heating in furnace at a temperature exceeding the given heat-treating temperature; c—heating with furnace; d—stepwise heating; t_s —temperature at surface of the heated part; t_c —temperature of its central portion.

heating up period. The thicker the part, the greater is the temperature differential from its exterior to its interior. Fig. 52 shows the experimentally established change in temperatures in the centre and at the surface of a forging 800 mm in diameter, made of steel of grade 40XII, heated in a furnace with a temperature of 950°.

Heating an article gives rise to its linear and volumetric expansion. A non uniform heating causes an unequal linear expansion. Let us suppose that a massive shaft is heated (Fig. 53): its exterior layers l after being heated up to the temperature t_1 , should expand and reach

the dimensions indicated by the dotted line (in a very exaggerated scale). The interior layers 2, being heated only to the temperature t_2 , also expand, but to a lesser extent, as $t_2 < t_1$. If these two layers 1 and 2 are not firmly united with each other but are like one cylinder 2 inserted in tube 1, then as a result of such a non-uniform heating an annular space should result between them. But since in reality layers 1 and 2 are firmly united, layer 2 inhibits the expansion of layer 1 which, in turn, compels layer 2 to expand to a greater degree

than could be caused by heating only. Consequently the interior layers 2 are under tension, and the exterior layers 1—under compression. In the article

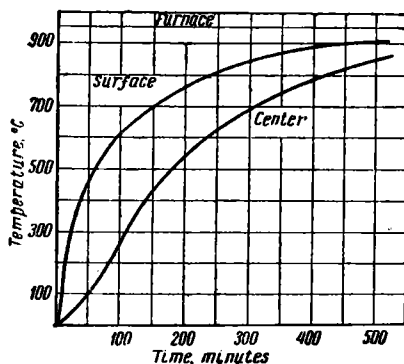


Fig. 52. Change in temperatures at the surface and in the centre of a forging made of steel of grade 40XH. The forging, 800 mm in diameter, is heated in furnace at a temperature of 950°C (A. A. Astafyev and K. A. Yermakov).

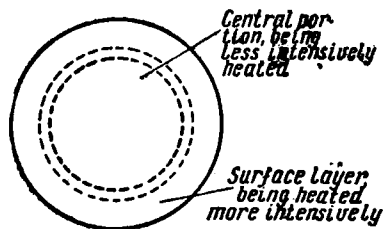


Fig. 53. Diagram showing the change in dimensions of various layers in a cylindrical forging being heated uniformly.

which is gradually and freely heated in a furnace and not subjected to any forces, *internal stresses* such as the compressive stresses in the exterior layers and tensile stresses in the central portion are set up.

This is a common phenomenon. The internal stresses are always developed where unequal temperatures arise in various parts of the article as a result of its non-uniform heating or cooling. When these internal stresses exceed the tensile strength of the material, cracks inevitably develop in the article.

So the heating of *massive articles* is always *non-uniform*, the degree of non-uniformity being the greater the heavier the article and the lower the thermal conductivity of the steel. The less uniform the heating, the higher the resulting internal stresses. But the thermal conductivity of the steel depends to a considerable extent upon its chemical composition, i. e., the higher the steel in carbon and alloying elements, the lower its thermal conductivity.

Grade of steel	Coefficient of thermal conductivity, Cal/metre per hour per degree Centigrade
10	75
25, 30, 35	70
45, 50, 55	62
35XM	38
30XH3 and 35XH3M	34
P18	21
1X13	20
1X18H9	14
1X41H14B2M	12

This is why, under other equal conditions, the heating of articles made from high-carbon and highly alloyed steels must be conducted at a slower rate than that of low-carbon and low-alloy steels.

As the temperature rises, the tensile strength and yield point decrease, as is evident from the following data for steel of grade 35:

Testing temperature, °C	Tensile strength	Yield point
20	55	33
100	52	31
200	59	31
300	59	21
400	51	19
500	37	15
600	20	8

From this one might suppose that the higher the heating temperature, the slower should be the rate of heating. In reality this is not so. The internal stresses are dangerous at temperatures between 500 and 600°, when steel has relatively high elastic properties and comparatively moderate ductility. At temperatures above 500 to 600°, steel almost entirely ceases to be elastic and becomes very ductile. For this reason the internal stresses developed in the steel at such temperatures are gradually relieved as a result of the local plastic deformations caused by them. The remaining internal stresses become negligible. Therefore, at temperatures above 500 to 600° steel articles may be heated faster, in practice at any heating rate which can be obtained in a given furnace with a certain definite heating efficiency.

In some cases the heating of massive articles is conducted in steps, i. e., the piece is heated in furnace up to some intermediate temperature, usually to between 500 and 600°, held at this temperature for several hours in order to equalise the temperature across the section, and then further heated at the fastest possible rate for the given furnace. In this case the time of heating from the temperature of the step to the final heat-treating temperature becomes shorter.

25. DETERMINATION OF THE HEATING TIME

The heating time can be determined either on the basis of a theoretical calculation or by means of an empirical formula derived from practical experience.

The differential equations of the heat transfer from the environment to the article to be heated are based on the theoretical calculations. These differential equations are resolved by means of special graphs. Determination of the time required for heating by theoretical

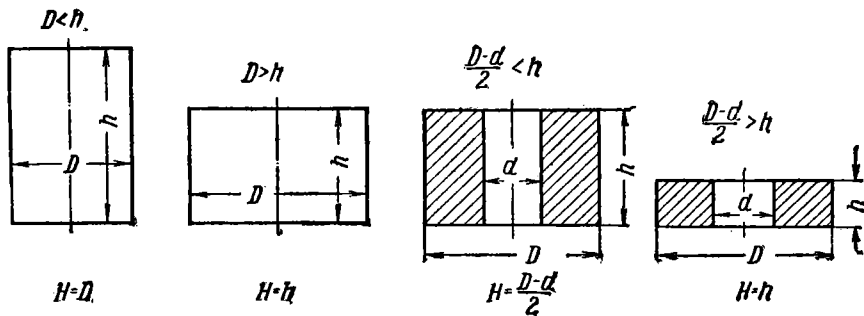


Fig. 54. Determination of the rated dimension H for parts of various shapes.

calculation is a cumbersome and complicated task, and the final results are only approximative. This is explained by the fact that the numerical values of many physical parameters included in the theoretical formulae are determined as yet rather inaccurately (especially for high temperatures). Therefore, when calculating by theoretical formulae a number of simplifying assumptions have to be made, and approximative values have to be adopted.

In practice, the empirical formulae are used in most cases for determining the time required for heating. This is accomplished on the basis of a gauged length H of the article. The minimum thickness of the piece (Fig. 54) is accepted as a gauged length H .

For articles of simple shape made of the carbon steels, which may be heated in a batch-type furnace at a constant temperature (700 to 900°) without preliminary heating, the time of heating for quenching (holding time included) can be determined as:

$$z_{\min} = 1.0 H_{\text{mm}}. \quad (1)$$

The heating time for tools made of carbon steel and for alloy steel articles can be defined as:

$$z_{\min} = 1.4 H_{\text{mm}}. \quad (2)$$

The time of preliminary heating for articles and tools of complicated shape which are previously heated in batch-type furnaces to a temperature of 500 to 600° is:

$$\text{for articles made of carbon steels } z_{\min} = 1.25 H_{\text{mm}}; \quad (3)$$

$$\text{for tools made of carbon steels } z_{\min} = 1.8 H_{\text{mm}}; \quad (4)$$

$$\text{for articles made of alloy steels } z_{\min} = 1.6 H_{\text{mm}}; \quad (5)$$

$$\text{for tools made of alloy steels } z_{\min} = 2.0 H_{\text{mm}}. \quad (6)$$

The time of heating from the temperature of preliminary heating (500 to 600°) to the final heat-treating temperature is determined by formulae, e. g.:

$$\text{for carbon steels } z_{\min} = 0.7 - 0.8 H_{\text{mm}}; \quad (7)$$

$$\text{for alloy steels } z_{\min} = 1.0 - 1.2 H_{\text{mm}}. \quad (8)$$

The time of heating and holding the articles when tempering them at elevated temperatures is determined by means of formulae (3) and (4).

Table 7 also may be used for determining the time required for heating parts made of the structural steels.

Table 7

Time of Heating and Holding Parts of Structural Steels
(according to data of "Giproavtoprom")

Size of cross-section in mm	Time in minutes							
	Quenching				Tempering			
	Reverberatory		Salt bath		Reverberatory		Salt bath	
	Heating	Holding	Heating	Holding	Heating	Holding	Heating	Holding
25	20	5	7	3	25	10	10	5
50	40	10	17	8	50	15	25	6
75	60	15	24	12	75	20	35	9
100	80	20	33	17	100	25	45	12
125	100	25	40	20	125	30	55	14
150	120	30	50	25	150	40	65	15
175	140	35	55	30	175	45	70	20
200	160	40	65	35	200	50	90	20

Note: For articles made of alloy steels the heating and holding time must be increased by 25 to 40 per cent.

The heating in salt baths is accomplished considerably faster than is possible in furnaces with air atmosphere, because the heat transfer from a liquid to a solid body occurs at a much more rapid rate. The time of heating the articles up to the temperature of molten salt may be determined from Tables 8 and 9.

Table 8

Time of Heating in Salt Baths

Relationship of the volume of the article to its surface area	Time of heating in minutes at temperature (°C)		
	850	950	1050
0.5	4	3	2
1.0	15	11	7
1.5	32	22	14

Table 9

Time of Heating in Saltpetrous Baths

Relationship of the volume of the article to its surface area	Time of heating in minutes at temperature (°C)		
	300	450	600
0.5	3.5	2.5	1.4
1.0	8	6	4
1.5	14	10	7

The time of heating in salt baths may also be determined by means of the empirical formulae, e. g.:

heating the parts of the carbon and alloy steels from room temperature to the heat-treating temperature:

$$z_{\min} = 0.35H_{\text{mm}}; \quad (9)$$

heating from a temperature of between 500 to 600° (the temperature of preliminary heating) to the heat-treating temperature:

$$\text{parts of carbon steels } z_{\min} = 0.15H_{\text{mm}}; \quad (10)$$

$$\text{parts of alloy steels } z_{\min} = 0.20H_{\text{mm}}. \quad (11)$$

The end of the actual heating process is indicated by a thermocouple in contact with the article to be heated (otherwise it would show the temperature in furnace, not that of the part), and by the colour of the heated part. If the article is heated in furnace at a temperature equal to that of the given heat-treating process, the period of proper heating ends when the colour of the article surface becomes similar to the colour of the furnace lining. As soon as this colour has been reached, the holding period begins, the time of which may be determined through the following equation:

$$z_{\min} = 0.5 - 1H_{\text{mm}}. \quad (12)$$

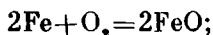
By simultaneous heating of a large number of parts in a single batch, when the formula (12) cannot be used, the holding time is

usually accepted to be from $\frac{1}{4}$ to $\frac{1}{5}$ of the time required specifically for heating. More accurate values, derived from experience, for the time of heating of the most important articles and tools will be given in the third part of this book.

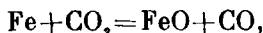
26. OXIDATION AND DECARBURISATION OF STEEL

When steel articles are heated, their outer layers react with the environment, i. e., with flue gases in reverberatory furnaces, with air in electric furnaces, and with fused salts in salt baths. As a result of this reaction two processes take place: 1) oxidation, and 2) decarburisation.

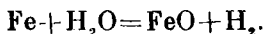
Oxidation of a steel is caused by oxygen:



carbon dioxide:



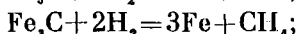
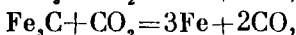
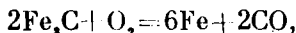
and water vapour:



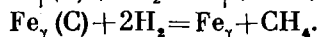
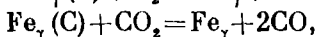
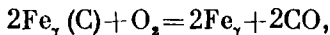
As a result of oxidation, an oxide film and then a thicker scale layer are formed at the surface of the steel. The higher the temperature and the longer the steel is kept at high temperatures, the more violently the oxidation takes place and the heavier is the layer of scale.

Decarburisation is a process of selective (preferential) oxidation of carbon; the carbon is oxidised, while iron remains non-oxidised. The decarburising process proceeds according to the following reactions:

by heating below the critical temperatures, when carbon exists in the form of cementite:



and by heating above the critical temperatures, when carbon is present in austenite:



Both these processes, i. e., oxidation and decarburisation usually proceed simultaneously. If the speeds of both processes are equal or if the rate of oxidation is faster than that of decarburisation, then the surface layer of the steel lying immediately beneath the scale layer is not decarburised and has a normal chemical composition. But if,

as usually happens, decarburisation progresses at a higher rate than oxidation, the surface layer of the steel loses its carbon and is composed only of the ferrite grains (Fig. 55).

Oxidation is an undesirable and, in many cases, harmful phenomenon, because a part of the heated metal is lost in the form of oxides (it is said to be "*burned*"). It is harmful when it is necessary to preserve the precise dimensions of articles being heat-treated. The loss of metal by burning inevitably diminishes the dimensions of the articles. If the thickness of the oxidised layer is less than the allowance for

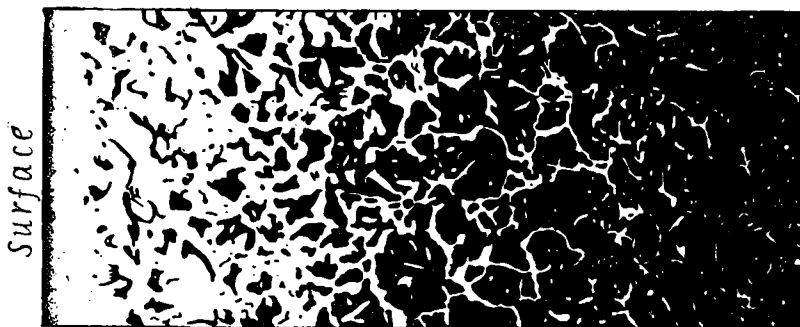


Fig. 55. Decarburised surface layer of a steel part, 100 $\times$.

machining it will be completely removed. But if the thickness of the oxidised layer is greater than the allowance for machining operation, then the part is considered irrevocably damaged.

Oxidation becomes very dangerous when steel is burnt. Burning of a steel manifests itself in the fact that a layer of partially oxidised steel lies under the layer of scale; oxygen has penetrated between the grains of the steel and has caused not only decarburisation but also oxidation of the iron itself, so that iron oxides have been formed at the grain boundaries (Fig. 56). It should be noted, however, that the burning of steel is completely excluded in modern heat-treating practice, because heating furnaces are now equipped with thermocouples and many of them even with automatic thermoregulators.

Decarburisation may also be a harmful phenomenon. If parts are to be machined after heat-treatment, a thin decarburised outer layer not exceeding the allowance for machining constitutes no danger because it will be completely removed in the machining operation. But if the rolled stock, articles or tools (springs, cold-drawn rods, steel ribbon, tools, etc.) are not subject to machining after heat-treatment, or if the thickness of the decarburised layer exceeds the machining allowance, decarburisation becomes very dangerous.

The danger of the decarburised surface layer present in steel articles or tools consists in their tensile strength, elastic properties, hardness, fatigue strength (endurance limit), and wear resistance being much lower than those of the more interior portions of the article. The danger of decarburisation becomes fully evident if it is remembered that according to the principles of strength of materials it is at the surface of articles that the torsional and flexural stresses are at their highest. Decarburisation is also very undesirable because it cannot be detected by visual inspection of parts and tools. It is an internal physical defect which can be revealed only by microscopic examination of the steel structure or through the measurement of its hardness.

Decarburisation depends upon the chemical composition of the steel. *Chromium makes steel less sensitive to decarburisation* because, first, it reduces the speed of carbon diffusion, and secondly, it favours the formation of a dense layer of scale which physically prevents the access of air to the surface of the article. On the other hand, *silicon, tungsten, vanadium and molybdenum increase the tendency of steel to decarburisation*.

Methods of determination of the decarburisation penetration are specified by the U.S.S.R. State Standard. To determine the depth of decarburisation a transverse microsection is prepared, the surface area of which should be perpendicular to the surface being investigated. This microsection is usually examined at a magnification of 100 diameters. If the article has been hardened, it must be annealed prior to the determination of thickness of the decarburised layer, the necessary precautions against eventual oxidation being naturally taken.

Two zones of decarburisation (Fig. 57) are distinguished: (a) zone of complete decarburisation and (b) zone of partial decarburisation.

If there are no other stipulations, the depth of decarburisation is defined as the sum total of the thicknesses of the zones of complete and partial decarburisation.

In highly alloyed tool steels (e. g., high-speed steels), the decarburised layer is not purely ferritic in structure. The decarburised layer in these steels consists of sorbite and primary carbides (as in the non-decarburised layers), but with a lesser amount of carbides. It is difficult to establish the difference in amount of carbides when examin-



Fig. 56. Structure of burned steel. Dark spots represent iron oxides, 100 $\times$.

ing the structure under microscope. V. D. Sadovsky has proposed the following method of determining the depth of decarburisation in high-speed steels.

This method is based upon the *M* point, indicating the beginning of the martensitic transformation, being dependent on the carbon level of steel (see Fig. 26) in the following manner: the less the percentage of carbon in the steel, the higher is the *M* point located.

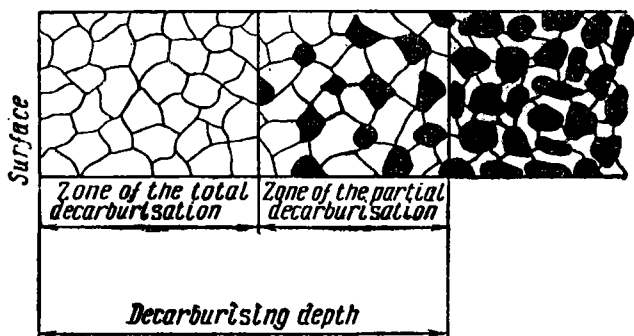


Fig. 57. Diagram illustrating what is meant by the term decarburising depth (State Standard).

The specimen of high-speed steel having a decarburised layer is quenched in oil or fused salt heated up to a temperature within less than 5 to 10 degrees of the *M* point for the given steel. For the outer decarburised layer, this quenching temperature of oil (or salt) is considerably lower than the *M* range. A marked decomposition of austenite into martensite occurs in the outer layer, while in its central layers this process just begins. To make the difference in structure between the surface and central portions more marked, the specimen is annealed at 580 to 600° immediately after quenching in oil (or fused salt). In this case the martensite of the decarburised layer partially transforms into troostite (this layer is found to be dark in the micro-section), while almost no changes will take place in the structure of the central portions, which will be of a light colour. Under such conditions the depth of decarburisation is fairly easily determined.

The U.S.S.R. State Standard of high-speed steels specifies the following procedure for determining the depth of decarburisation in these steels, based upon the method of V. D. Sadovsky:

1) A sample of dimensions shown by Fig. 58 is cut off from the rod stock.

2) The sample is heated in a properly deoxidised salt bath of barium chloride up to 1270 to 1280° (steel of grade P18) or to 1230 to

1240° (steel of grade P9), and kept at these temperatures for a period of 2 minutes from the moment of its being immersed in the bath.

3) The sample to be quenched in oil (or fused salt) is heated up to 160 to 180° and held at this temperature for 40 minutes.

4) After being withdrawn from the oil (or fused salt), the sample is immediately transferred into a salt bath heated up to 580 to 600° and held in it for a period of 10 minutes.

5) A microsection is prepared and etched with a 4 per cent solution of nitric acid.

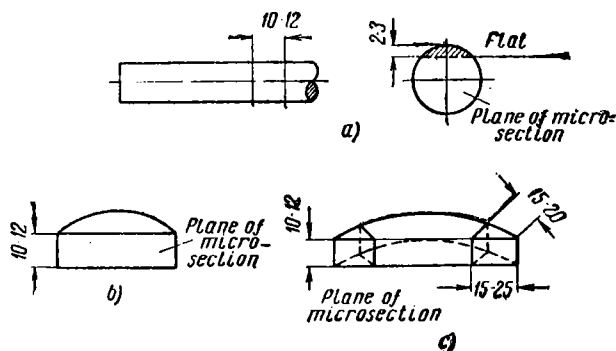


Fig. 58. Shapes of the specimens for determining the decarburising penetration in rolled stock made in high-speed steels (according to V. D. Sadovsky's method).

a—for diameter up to 25 mm; b—for diameter from 25 to 40 mm; c—for diameter over 40 mm.

If decarburisation is found in such areas of the microsection which do not correspond to the surface of the rod or tool, then the barium chloride bath itself is a cause of decarburisation. To reveal this a part of the surface layer is removed on the samples made from a rod stock of small section (Fig. 58, a), thereby preparing on them a sort of flat surface. When decarburisation is found in those areas of the microsection which do not correspond to the surface of the rod stock, then it is impossible to draw any definite conclusions about the depth of decarburisation for the steel under examination.

The U.S.S.R. State Standard of high-speed steels specifies the allowed thickness of decarburised layer in rolled stock as follows:

Dimension of rolled stock in mm	Depth of decarburisation in mm
5-15	0.40
15-30	0.50
30-50	0.70
50-70	0.80
70-80	1.00
80-100	1.35

27. MEANS OF PROTECTION OF STEEL FROM OXIDATION AND DECARBURISATION

To protect a steel against oxidation and decarburisation various means, from the simplest and most elementary to relatively complex and costly ones, are employed.

They may be divided into two main groups:

- 1) Isolation of parts to be heated from the furnace atmosphere.
- 2) Partial limiting or complete removal of oxygen and other oxidising gases (carbon dioxide, water vapour) from the furnace atmosphere.

Isolation of small parts from contact with the furnace atmosphere is usually effected by charging them into a pot or box containing iron chips and cuttings (swarf) or used carburiser. This method gives fairly good results in the case of annealing, but is somewhat less effective for hardening because a part of the iron chips or carbonaceous matter inevitably goes with the articles into the quench tank and therefore pollutes it.

For heating hammer dies, the following procedure is usually accepted: the hammer die is placed impression downwards on a metal base, the surface of which is covered with a thin layer of crushed charcoal or used carburising compound. The clearance between the die and base is sealed up by fireclay mixed with 10 to 20 per cent of ground asbestos.

For annealing thin rods, the adopted procedure is often that they are inserted into tubes, the holes of which are sealed up with fireclay.

In reverberatory furnaces the *limiting of content of oxygen and other oxidising gases in their atmospheres* is effected by the creation of a neutral or reducing atmosphere in the furnace chamber by regulating the combustion conditions. The appearance and colour of the flame may be considered as indicative of the character of the furnace atmosphere. When the atmosphere is oxidising, the flame is luminous and transparent; when reducing, it is dark and non-luminous. When there is a neutral atmosphere, the flame is half-transparent, white-coloured and with violet strips. It should be noted here that the reducing furnace atmosphere does actually prevent oxidation of the metal. As to decarburisation, it takes place under reducing atmosphere, too, being caused by the hydrogen which is always contained in such an atmosphere (see the decarburising reactions above).

In muffle and electric furnaces a slightly reducing atmosphere is created with charcoal (preferably of birch) being charged on the bottom of the furnace near to its doors; on burning this gives rise to the formation of carbon monoxide (CO). Such is the adopted procedure, for instance, in heating springs for hardening.

However, all these methods are found to be insufficiently effective, and in cases where large amounts of metal have to be heated (e. g., annealing of rods, wire and strip) they are simply impracticable. In such cases *heating in furnaces with a protective gas atmosphere* is absolutely indispensable and gives excellent results. In this method of heating some protective gas is used to supplant the air and to fill the entire heating chamber of the furnace. To exclude air infiltration the protective gas is present in the furnace under a slightly positive pressure. The protective gas atmosphere is most easily created in muffle and electric furnaces. In practice such an atmosphere cannot be created in the reverberatory furnaces.

The articles heated in a furnace with a protective gas atmosphere retain a completely bright surface after heating (so-called *bright heating*) or are only slightly tarnished (so-called *clean heating*).

The main types of industrial protective atmospheres used in heat-treatment of steel are: dissociated ammonia; industrial gases (natural, coke oven, lighting, etc.); producer gas obtained from charcoal; gases produced through pyrolysis (decomposition in heating) or cracking (decomposition in heating under high pressure) of some gas (e. g., natural gas), or some liquid rich in hydrocarbons (kerosene, fuel oil or naphtha residue, oil). These atmospheres cause an increase of the carbon content of the steel and are used for carburising steel parts. They are also employed as protective atmospheres in the bright annealing of high-carbon steels.

Under otherwise equal conditions the heating of the metal in salt baths is accompanied by its decarburising to a considerably less degree. This fact constitutes one of the great advantages of heating in salt baths. But some degree of decarburising may occur even in these conditions. The barium chloride baths employed in heating tools of high-speed steels for hardening (at 1250 to 1300°), possesses especially marked decarburising power. In these baths a gradual accumulation of the iron oxides takes place and causes the decarburisation of the metal.

The surface of tools to be heated in a barium chloride bath should be covered with a borax in order to prevent the decarburisation of the steel. This is done as follows: the tool is heated up to 800 to 850° and then sprinkled with borax, which melts and covers the tool surface with a thin layer. A simpler procedure is to prepare a hot saturated borax solution and to immerse the cold or (better) slightly warmed tool in it. When the water has been evaporated the surface of the tool is covered with a thin uniform deposit of borax. When heated in the bath it will melt and cover the tool with a thin layer.

It is still better to remove the cause of the bath decarburising action. This is achieved by periodic deoxidation of the bath, carried out once or twice a shift. For this purpose a small amount of finely

crushed ferrosilicon (1 to 1.5 per cent by weight to the total amount of salt in the bath) is added to the intensively heated bath (preferably the 75 per cent ferrosilicon), and the fused salt in it is then agitated. The heating of tools in the bath begins after it has been allowed to settle for half an hour. V. D. Sadovsky's method, described in the previous section, may be used to verify whether the barium chloride bath exerts a decarburising action upon the metal. For this purpose only samples without a decarburised layer must be selected.

28. TECHNOLOGICAL CHARACTERISTICS OF HEAT-TREATING FURNACES

Furnaces used in heat-treating practice differ widely according to their purposes, designs, methods of heating and charging.

Heat-treating furnaces may be classified on the basis of the following main technological characteristics:

- 1) according to use;
- 2) according to type of work;
- 3) according to source of heat;
- 4) according to working environment.

As to use, heat-treating furnaces are divided into annealing, hardening, tempering, carburising furnaces, etc. Some furnaces are of a universal kind, i. e., they are suitable for many types of heat-treating operations, if not all. For example, batch-type furnaces, also called in-and-out furnaces, which are extensively used in heat-treating shops, are equally fit for annealing, high tempering and pack carburising, as well as for heating metal for its normalising and hardening.

As to the type of work, heat-treating furnaces are divided into batch-type furnaces and continuous type furnaces. The batch-type furnaces operate as follows: a batch of parts is charged into the furnace, heated to some given temperature, held for some period of time and subjected to some heat-treating operation (annealing, heating for normalising or hardening); and then the whole batch of parts is withdrawn from the furnace. Subsequently the next batch of parts is charged into the furnace, undergoes the heat-treatment, etc. One batch of parts (one charge) differs from another in weight, size, grade of steel and type of heat-treating required. Consequently, the batches differ from each other both in heating temperature and in the time for which they are kept in the furnace. This is exactly the pattern of operation of the batch-type heat-treating furnaces, which are the most typical for single piece and small lot types of production. Batch-type furnaces are usually installed in heat-treating, tool-making and maintenance shops of machine-building plants with wide nomenclature but a relatively limited amount of parts and tools to be heat-treated. But,

it should be noted that the great variety of heat-treating operations performed in batch-type furnaces adversely affects the efficiency of their work, i. e., the productivity is lowered and the consumption of fuel or electric power is increased. Therefore, if there is a number of batch-type furnaces in the shop their work should be *specialised* as far as possible, so that one set of furnaces is used for annealing, another for normalising or hardening, still another for high tempering, etc. On the other hand, when the charges are small (or the amount of work is limited) it is advisable to operate only one or two furnaces with various heat-treating operations grouped in time.

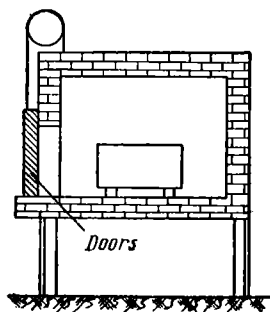


Fig. 59. Box-type furnace.

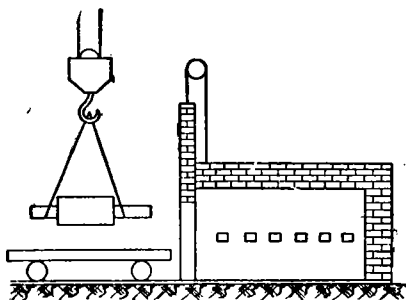


Fig. 60. Box-type furnace with movable bottom (car-bottom furnace).

Box-type furnaces with side loading and unloading are employed for heat-treatment of small parts (Fig. 59). Charging in such furnaces is usually made manually. To load and unload bigger articles (e. g., dies, carburising pots) use is made of the simplest mechanical means, e. g., single-girder overhead travelling cranes, monorail carriers or telfers with electric hoists, etc.

Box-type furnaces with a movable bottom (with the hearth mounted on wheels to permit it to be removed for loading and unloading) are used for heat-treatment of large parts (Fig. 60). Before loading the bottom is removed from the furnace and the part to be heat-treated is placed on it by means of an overhead travelling crane; the bottom is then replaced in the furnace. The unloading of the parts from the hearth after heat-treatment is effected in the same manner.

Heating of long parts of large size is best carried out in pit furnaces in a suspended (vertical) position, for the heating of such parts in a horizontal position brings about warping, caused by their own weight (Fig. 61). Pit furnaces may also be loaded with small parts in special charging baskets.

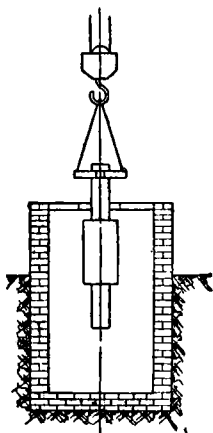


Fig. 61. Pit-type furnace.

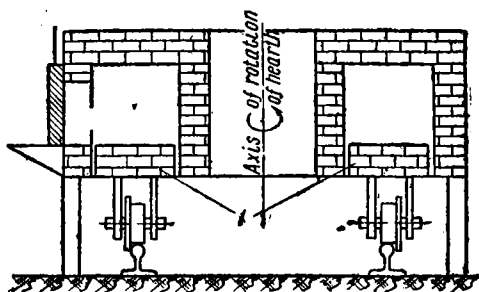


Fig. 62. Rotary-hearth furnace.

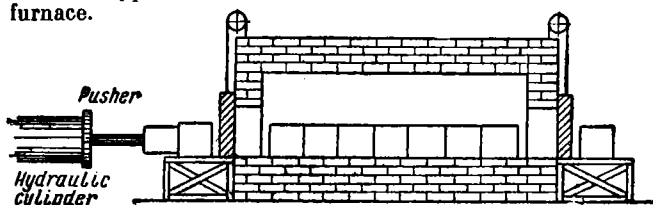


Fig. 63. Pusher-type furnace.

Large parts are also heat-treated in pit furnaces with removable covers. This type of furnace includes, in particular, soaking-pit furnaces intended for heating ingots.

In cases where the shop floor space is insufficient to install furnaces with a movable bottom, elevator-type furnaces prove to be suitable, the heating chambers of which are suspended at a height of several metres, while the hearths with parts loaded upon them are raised into the furnace chambers by means of hydraulic hoists. Elevator-type furnaces are employed in particular for annealing transformer sheet steel and castings of malleable iron. These furnaces are suitable for creating a protective gas atmosphere, the sealing of their heating chambers being easily effected with sand seals.

Continuous heat-treating furnaces always operate at permanent temperature conditions. They are always used to heat the same parts, made of the same grade of steel and subjected to the same heat-treating operations, e. g., only to annealing, to hardening, or to tempering. These furnaces are strictly *specialised*. Continuous furnaces are

typical of plants working on a mass production scale. In continuous furnaces the parts to be heat-treated are charged at one end of the furnace and discharged at the other end. The parts are moved through the furnace, as a rule, by mechanical means. Mechanisation of the movement of parts through the working chamber may be effected in various manners. The continuous furnaces most commonly used are as follows:

1) Rotary-hearth furnace (Fig. 62) with a ring-like bottom slowly rotating around the vertical axis; they do not require much floor space, nor do they require boxes made of non-scaling steels.

2) Pusher furnace (Fig. 63), in which the work is pushed through the working chamber by means of a pusher with hydraulic or mechanical drive; the pusher is designed to push all the parts periodically through the furnace over a certain predetermined distance; large pieces are loaded directly on the hearth, and small ones on trays; pusher furnaces are used for annealing and tempering, as well as for hardening and normalising.

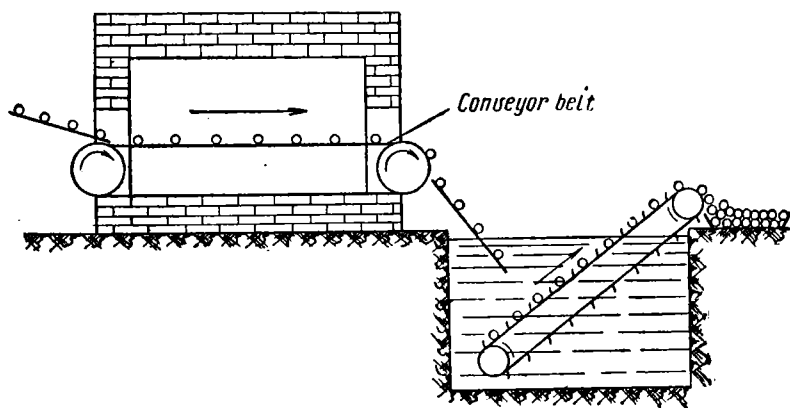


Fig. 64. Conveyor furnace.

3) Conveyor furnace (Fig. 64), in which the belt conveyor continuously moves at a very slow rate; the links of the belt conveyor are cast of non-scaling steel; parts to be treated are directly placed on the belt; as soon as they have reached the other end of the furnace, they are poured into boxes or dropped into the quenching tank. Conveyor furnaces are most frequently employed for hardening or tempering.

4) Tunnel furnace (Fig. 65), in which the stock to be heated (parts, stacks or piles of sheets, etc.) is placed upon a number of cars, which are then periodically pushed or pulled slowly through the furnace at a

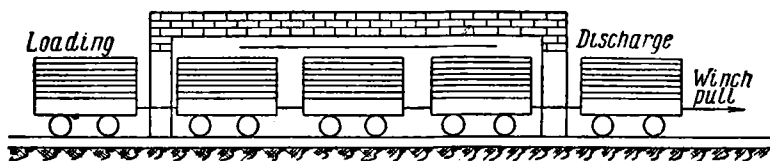


Fig. 65. Tunnel furnace.

distance equal to the length of a car by means of a moving device. Tunnel furnaces are usually used for annealing.

5) Rotary furnaces (Fig. 66) slowly rotating around the horizontal axis, through which the parts to be treated are moved by means of ribs placed at right angles to the furnace axis and forming a continuous spiral line (screw conveyor). Rotary furnaces are used for heating small parts for hardening and tempering, as well as for gas carburising. Hot gases are made to circulate around the articles to be heated through continuous rotation of the furnace.

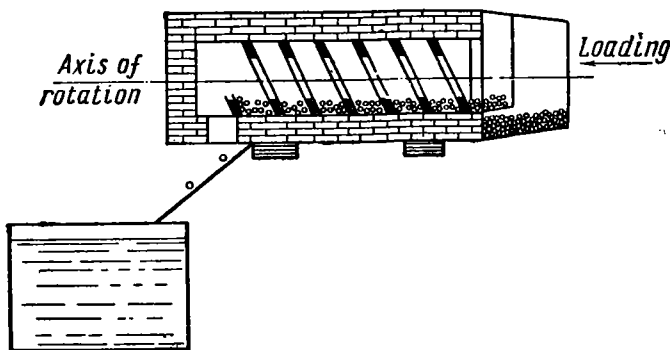


Fig. 66. Rotary furnace.

Apart from the continuous furnaces of the types just described, there are also roller-hearth furnaces, walking beam furnaces, vibrating hearth furnaces, etc. (which are employed considerably less often).

Heat-treating furnaces are classified as follows, according to the source of heat: 1) *reverberatory* furnaces, which are heated through combustion of fuel, such as solid fuel (coke or coal), liquid fuel (fuel oil), gaseous fuel (natural gas, producer gas, etc.) and 2) *electric* furnaces.

In modern heat-treating shops, *coke or coal-fired* furnaces are now hardly ever used. They differ from oil-fired and gas-fired furnaces in

their considerably more complicated design. In these furnaces precise control of the temperature conditions can be achieved only with great difficulty. Automation cannot be introduced at all.

Oil-fired furnaces are better in the following respects: their design is simpler; there are no ashes remaining after the combustion of fuel oil; temperature control is effected with great accuracy. However, the organisation of fuel-oil preparation plant in the shop presents fairly difficult problems, e. g., the fuel oil should be stored in a separate building located, as a rule, outside the shop building; in winter-time, the fuel oil thickens, so that the oil pipe lines have to be specially warmed up by steam.

Of all reverberatory furnaces the *gas-fired* ones are the best and most economical. Being of simple design, they are easily controlled, so that they do not present substantial difficulties in running. Gas is led into the furnaces through the gas mains from the central plant gas-generating station or from outside sources of gas. The use of producer gas for firing the heat-treating furnaces is more appropriate from the technical point of view than a direct combustion of solid fuel, being at the same time more advisable from the economic viewpoint than combustion of fuel oil. Gas-fired furnaces are highly perfected, being inferior only to electric furnaces.

Heating furnaces with *electric current* is the most perfect means of power use. Electric heat-treating furnaces are very simple in design; they have neither combustion chamber, gas ducts, nor stack flues. The temperature in electric furnaces is controlled very readily. In addition, the temperature control may be wholly and relatively easily effected by means of automation. In electric furnaces the heat utilisation factor is the highest because there are no heat losses with exit gases. The efficiency of batch-type electric furnaces intended for normalising and hardening ranges from 65 to 75 per cent, while that of reverberatory furnaces is from 15 to 20 per cent. The efficiencies of continuous heat-treating furnaces are 70 to 80 per cent and 14 to 25 per cent, respectively. Conditions of work in running electric furnaces are considerably more hygienic than those of reverberatory furnaces.

But, if almost all heat-treating furnaces, from the smallest to the largest, can be heated through gas firing, electric heating can be used today only for comparatively small furnaces. Gradually, however, as the electrification of the U.S.S.R. progresses, electric heat-treating furnaces will increasingly replace all other types.

The *working media* in which the heating of articles may be performed are as follows:

- air (in most heat-treating furnaces);
- protective gas atmospheres;

special gas atmospheres (used in gas chemical-heat treatment);
fused salts (salt baths);
mineral oils (oil baths for tempering);
powdered matters (pack carburising, calorising, etc.).

In reverberatory furnaces, where protective or special atmospheres are not used, heat is transmitted from the hot gases to the articles by radiation and directly through convection, i. e., through gas currents washing the articles.

In electric furnaces with metallic resistance heating elements heat transmission is effected only through radiation. From this two conclusions may be drawn: first, electric furnaces heat somewhat slower than reverberatory furnaces; secondly, at relatively low temperatures, e. g., in high tempering, heating in electric furnaces progresses slowly and non-uniformly. It should be noted here that according to the Stefan-Boltzman Law the transmission of heat by radiation is proportional to the difference of the absolute temperature fourth powers for the heater and the body to be heated. Therefore in such furnaces the hot gases should be made to circulate around the articles to be heated in order to intensify the heat transmission by convection. In electrically heated annealing furnaces a fan is provided in the roof of each furnace, producing the forced circulation of hot gases for this purpose.

Muffle furnaces are used for bright heating of articles in protective atmospheres or for such heat-treating operations as carburising, nitriding and cyaniding (the so-called chemical-heat treatment). There are two main types of batch-type muffle furnaces which differ from each other in design. In pit-type furnaces (Fig. 67) the muffle has the form of a cylindrical retort made of non-scaling steel, which is placed in a vertical position inside the furnace. The articles to be treated are loaded into this retort, its cover is hermetically sealed and some proper gas (protective gas or gas intended for the chemical-heat treatment) is let in. Heating is effected by hot furnace gases, which are made to circulate through the ring-like space between the interior furnace wall and the exterior muffle wall. In electrically heated muffle furnaces the heating elements are placed in the same ring-like space. Such are the pit-type furnaces designed for bright annealing, bright normalising, gas carburising and nitriding. Gas carburising of small parts is carried out in muffle furnaces with a cylindrical retort revolving around the horizontal axis.

In muffle bell furnaces (Fig. 68) the muffle chamber has the form of a removable bell, which is lifted and moved from base to base with the material to be treated loaded on it. The sealing is obtained through the use of sand seals.

The weak point of muffle furnaces is the short service life of the muffle. Therefore gas-fired furnaces with radiant tubes have been

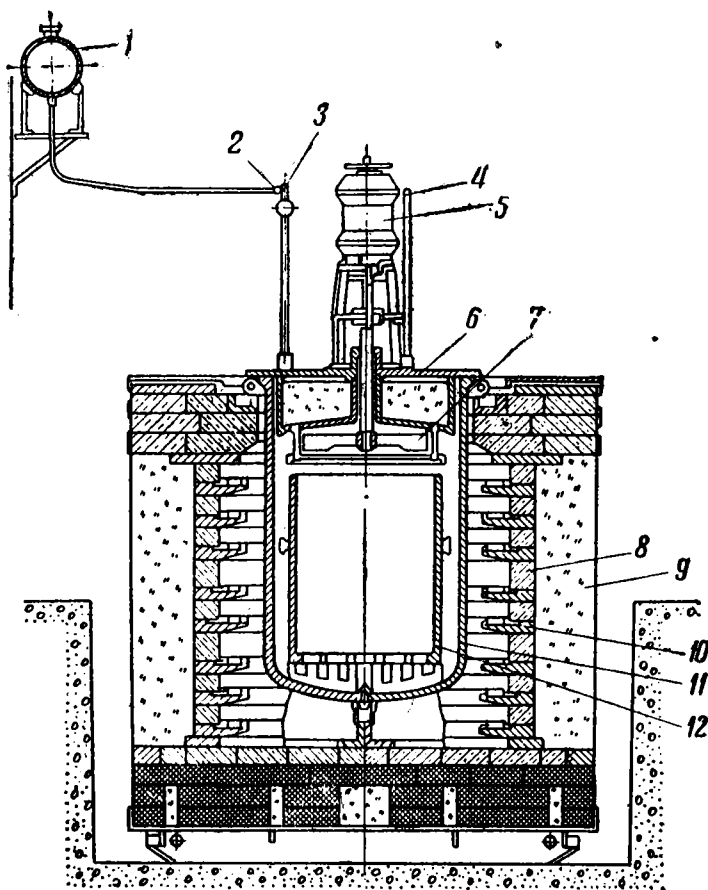


Fig. 67. Electrically-heated pit-type furnace for gas carburising.
 1—tank for hydrocarboneous liquid; 2—drip nozzle; 3—valve; 4—outlet pipe for exit gases from the furnace chamber; 5—motor for fan; 6—cover of the furnace; 7—fan; 8—fire-brick lining of the furnace; 9—heat-insulating powder; 10—electrical heaters; 11—retort; 12—basket for loading small parts into the pit furnace.

developed, the application of which is now increasing. In these furnaces the heating is accomplished by the burning of gas admitted to the radiant tubes, i. e., the flame is forced through the heating or muffle tubes. There is thus no necessity for a muffle.

Salt baths are used for heating small parts (Fig. 69). In these furnaces the working medium consists of fused salt inside a steel or cast-

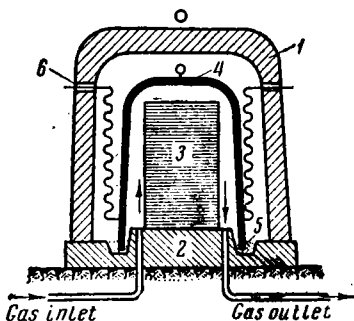


Fig. 68. Muffle bell furnace for bright annealing.

1—cover; 2—base; 3—parts to be treated; 4—bell; 5—sand seal; 6—electrical heaters.

iron pot. This pot is heated and maintained at a proper temperature either by combustion of a fuel (fuel oil or gas) or by electrical resistance. Apart from salt baths with external, or indirect, heating, there are others with internal heating, the so-called direct heated salt baths, the heating of which is accomplished through electrodes (Fig. 70). In these bath furnaces a fused salt is at the same time a working medium and the heating element, for it serves as a conductor or resistor in which heat is developed. Electrode salt baths have a higher heating efficiency, so that higher temperatures can be achieved in them. Therefore electrode salt baths are always used to heat tools of high-speed steel for hardening.

The composition of salt mixtures to be used in salt baths is shown in Table 10.

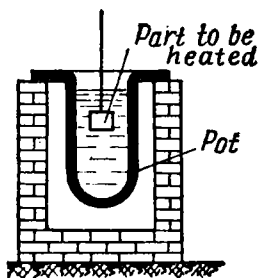


Fig. 69. Salt bath.

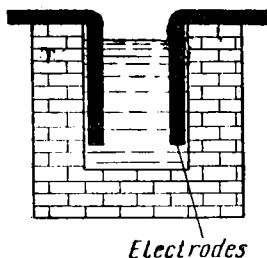


Fig. 70. Electrically-heated salt bath.

Low tempering (up to 250°) and artificial ageing must be carried out only in electrically heated oil baths, for a uniform warming cannot be achieved by heating in air.

Table 10

**Salts and Salt Mixtures Most Frequently Used
for Salt Baths**

Main purpose of the salt bath	Salts and salt mixtures		Temperature, °C	
	Name of salt	per cent	melting	working
Saltpetrous bath for tempering springs and articles	Sodium saltpeter	55	218	230-550
	Potassium saltpeter	45		
	Sodium saltpeter	45	317	325-600
	Potassium saltpeter	55		
Salt baths for heating carbon and low-alloy steels to be hardened	Sodium saltpeter	100	317	325-600
	Potassium saltpeter	100	337	350-600
	Sodium chloride	35	620	650-900
	Calcined soda	65		
Salt baths for heating highly alloyed steels (X12, X12M and others) to be hardened	Sodium chloride	22	654	675-900
	Barium chloride	78		
	Sodium chloride	100	810	850-1100
	Sodium chloride	100	960	1100-1350
Alkaline baths for bright hardening	Caustic soda	100	322	350-700
	Caustic potash	100	360	400-650
	Caustic soda	37	159	180-350
	Caustic potash	63		

To prepare the oil baths the following sorts of oil are usually employed:

Name of oil	Flash point of oil, °C (not less than)
Cylinder oil 6	300
Vapour	310
Vapour C	310
Viscosine	240

Chapter VI

ANNEALING AND NORMALISING

29. PURPOSES OF ANNEALING AND NORMALISING

The various purposes of annealing and normalising steel are listed as follows:

- 1) to improve all mechanical properties;
- 2) to improve the machinability;
- 3) to increase ductility, particularly to restore the normal condition of steel after cold working;
- 4) to remove chemical non-uniformity;
- 5) to alter the microstructure and develop a structure more desirable for hardening;
- 6) to relieve internal stresses.

So the problems of annealing and normalising which have to be worked out under manufacturing conditions are fairly numerous and varied. The various types of annealing that are to be used to resolve the above-listed problems are based upon various physical phenomena. Annealing based upon phase recrystallisation processes is given different names, i. e., phase recrystallisation, annealing of the second kind, or simply annealing. For the sake of brevity we shall term it *phase annealing*. Normalising is also based upon phase recrystallisation. An annealing using the recrystallisation phenomenon is called *recrystallisation annealing*. An annealing performed to even out the chemical non-uniformity of the steel is called diffusion or *homogenising annealing*, or *homogeneisation*. *Spheroidising* or *softening* annealing is a term applied when the annealing is based upon the coagulating and spheroidising processes. Finally, an annealing, the main purpose of which is to relieve internal stresses and which is based upon the relaxation process (plastic deformations in microvolumes as a result of internal stresses) has not been given a specific name, and is called *annealing for removing or relieving internal stresses*.

Table 11

**Purposes of Various Types of Annealing and Normalising
of Plain Carbon and Low-Alloy Steels**

Purpose of annealing or normalising	Blanks, billets and parts	Type of annealing or normalising operation
To improve the mechanical properties in general	Steel castings Parts made of structural steels	Phase annealing Normalising
To improve the machinability of the steel	Low-carbon steels Medium-carbon structural steels Tool steels	Normalising Phase annealing or normalising, followed by high tempering Spheroidising (to obtain pearlite in globular condition)
To increase ductility	Low-carbon steels Medium- and high-carbon steels	Recrystallisation annealing Annealing to obtain pearlite in globular condition, or normalising
To remove chemical non-uniformity	Castings and ingots	Homogenising annealing
As a treatment preliminary to hardening, in order to develop a more desirable structure	Parts and tools	Normalising
To relieve internal stresses	Castings and forgings Welded steel structures	Phase annealing Low-temperature annealing to relieve internal stresses

Table 11 shows which type of annealing or normalising is to be used to achieve some particular objective among those listed above.

Fig. 71 represents graphs showing various types of annealing and normalising. From these graphs it is clear that phase annealing, normalising and homogeneisation are carried out at temperatures above

the critical ones (at least, over the A_1 point), whereas recrystallisation annealing, spheroidising and annealing for relieving internal stresses are done at temperatures below the A_1 point. Strictly speaking, recrystallisation annealing and annealing for relieving internal stresses proceed independently of phase transformations, and therefore they may be carried out at temperatures both above and below the A_1 point; in practice, however, these heat-treating operations are usually accomplished at temperatures under the A_1 point.

The common feature peculiar to all types of annealing is the approach to the phase and structural equilibria. It is in this that any

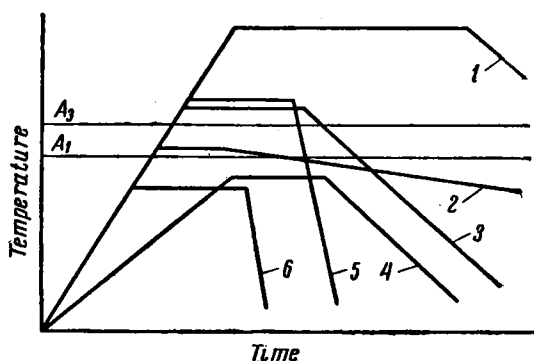


Fig. 71. Diagram representing the various types of annealing operations.

1—homogenising; 2—annealing to obtain globular pearlite; 3—phase annealing; 4—annealing to relieve internal stresses; 5—normalising; 6—recrystallisation annealing.

type of annealing differs from hardening, in which more or less considerable deviations from the phase and structural equilibria are brought about. From this standpoint normalising resembles hardening, i. e., in normalising there occurs some deviation from the structural equilibrium. From the industrial point of view, however, normalising is nearer to annealing, therefore normalising will here be considered with annealing.

The fact that several types of annealing are carried out at temperatures below the A_1 point is not sufficient reason to identify them with tempering, which is also accomplished always at temperatures below the A_1 . The initial condition of a steel to be subjected to high tempering is, as a rule, martensite, i. e., a structure with unstable phase composition which changes in tempering, while in low-temperature annealing (recrystallisation and spheroidising) the phase composition of steel does not change at all.

30. ANNEALING FOR GENERAL INCREASE IN MECHANICAL PROPERTIES

The mechanical properties of steel can be improved by means of annealing, normalising and hardening, followed by a high tempering. Hardening combined with high tempering make up a complicated heat-treating operation which is generally referred to as *improvement*.

In all three heat-treating processes mentioned above the mechanical properties of steel are improved, but they improve to different extents, as is clear from the data shown in Table 12 (S. K. Ilyinsky, A. P. Gulyaev, and A. P. Sanina).

Table 12

Mechanical Properties of Steel of Grade 40X Subjected to Various Types of Heat-Treating Operations

Type of heat-treatment	Tensile strength	Yield strength	Elongation	Contraction	Impact value
Annealing	65.6	36.4	21.0	53.5	5.6
Normalising	75.4	45.0	20.9	56.0	7.8
Improvement (hardening with subsequent high tempering)	87.6	76.5	22.5	67.5	16.8

The best results from the point of view of mechanical properties are obtained in hardening with subsequent high tempering, somewhat lower results in normalising, and still lower results in annealing. However, it must not be concluded from this that only the above-mentioned kind of heat-treatment, i. e., hardening with subsequent high tempering, should always be carried out to improve the mechanical properties of a steel. Of all three types of heat-treating processes this is both the most complicated and the most expensive. Therefore only high-duty machine parts, particularly those made of alloy steels, when the highest mechanical properties are required to meet the demands of working conditions, e. g., crank-shafts, gears, springs, etc., are subjected to hardening with subsequent high tempering. For most types of moderately loaded machine parts made of plain carbon steels annealing and normalising will be sufficient.

Of these two types of heat-treating operations, i. e., *annealing* and *normalising*, the latter is to be preferred. Normalising has the following advantages over annealing: First, in normalising the mechanical properties of a steel somewhat improve (see Table 12), e. g., forged shafts of plain carbon steel are therefore most often subjected to normalising but not to annealing. Secondly, normalising has sub-

stantial advantages over annealing from the process point of view, i. e., in annealing the part to be treated is cooled with a furnace, whereas in normalising parts are withdrawn from the furnace and cooled in still air at room temperature, while the furnace may be employed for heat-treating the subsequent batch of parts. In annealing the furnace is cooled down to comparatively low temperatures, and should then be heated again. For this reason the cycle of heat-treating operation in annealing is longer than that in normalising and consumption of fuel or electric power is also greater.

On the other hand, annealing has some advantages over normalising, e. g., steels containing over 0.3 to 0.4 per cent carbon have better machinability in an annealed condition, and during annealing the internal stresses are relieved to a greater degree than in normalising.

Thus, it may be concluded that if the improvement of mechanical properties is the main objective of the heat-treating operation, then normalising should generally be performed. To reduce the increased hardness of normalised steel, which is especially important for steels with higher carbon content, and to relieve internal stresses, normalising is followed by high tempering (at 600 to 680°C). Such heat-treating procedure is used, for instance, in cases of forged shafts and rolled stock.

However, in cases where the main purpose of the heat-treating operation is not so much the improvement of mechanical properties as that of machinability and particularly the considerable relief of internal stresses, annealing is performed (e. g., annealing is used with steel castings). In many cases annealing may be substituted for normalising with subsequent high tempering. Parts of complicated shapes are the only exception in this respect.

The improvement of the mechanical properties of steel after annealing and normalising is explained first of all by the formation of fine grain structure as a result of phase recrystallisation. In heating the hypoeutectoid steel above the temperature corresponding to the A_1 point its initial structure, consisting of ferritic and pearlitic grains, is *totally replaced* by the structure consisting of grains of austenite. These, like any other newly formed grains, are originally fine, and if the steel is not superheated, substantial growth of grain will not occur (see Fig. 14). With subsequent cooling in the critical range A_1 - A_c , the fine austenitic grains transform into equally fine grains of ferrite, while at temperatures below the A_1 point they become fine grains of pearlite. The phase composition of a steel both before and after annealing and normalising is absolutely the same, while the structure of steel prior to annealing or normalising may be both fine grained and coarse grained, being only fine grained after annealing and normalising (if there is no superheating).

The steel is annealed or normalised for the fine grain structure to be developed only when its structure has been coarse grained. In this respect the possibilities offered by annealing and normalising operations are enormous (Fig. 72). It should be noted here, by the way, that a very limited number of alloys possess such a property. Besides the steels many alloys are capable of being hardened, e. g., most aluminium alloys, several types of bronze, etc. But not one of these alloys is capable of changing its grain size as a result of the annealing or normalising operation.

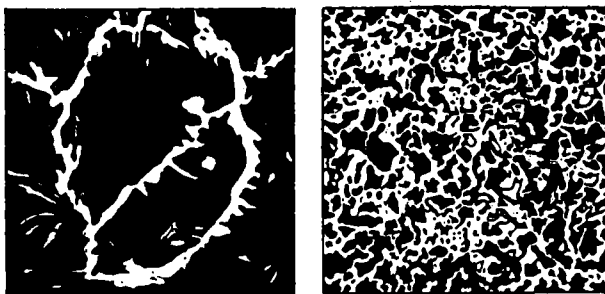


Fig. 72. Microstructure of plain carbon steel of grade 35 before (to the left) and after (to the right) annealing, 100 $\times$.

As is evident, the normal annealing temperature for hypoeutectoid steels must somewhat exceed the critical temperature corresponding to the A_1 point:

$$t_{\text{anneal}} = A_1 + (30-50^\circ\text{C}). \quad (13)$$

A small excess of temperature above the A_1 point is provided to equalise the chemical composition of austenite.

In normalising a somewhat higher temperature (by 30 to 50 $^\circ\text{C}$) is adopted than that computed by means of the above formula. This is made with a view to further increasing the uniformity of austenite and its stability under supercooling. As is already well known, more drastically supercooled austenite decomposes into a more dispersed aggregate made up of ferrite and cementite. But a more dispersed conglomerate possesses higher mechanical properties. Hence the increase in mechanical properties of a steel after normalising has two causes: 1) the finer grain size; 2) the more dispersed structure of the ferrite-cementite conglomerate.

Table 13 shows the annealing and normalising temperatures for carbon structural steels.

Table 13

Annealing and Normalising Temperatures (in °C)
for Carbon Structural Steels

Grade of steel	A_1	Temperature, °C	
		Annealing	Normalising
08	890	Not to be annealed	920-970
Ст. 1, Ст. 2, 10	880	• • • •	910-960
Ст. 3, 15	865	• • • •	900-950
Ст. 4, 20	850	• • • •	880-930
25	840	• • • •	870-920
Ст. 5, 30	825	850-870	850-900
35	810	840-860	840-890
40	800	830-850	830-880
Ст. 6, 45	785	820-840	820-870
50	770	800-820	800-850
Ст. 7, 55	765		
60	760		
65	750	790-810	790-840

Underheating of steel, i. e., heating it to temperatures below the A_1 point in annealing or normalising, results in incomplete annealing or incomplete normalising.

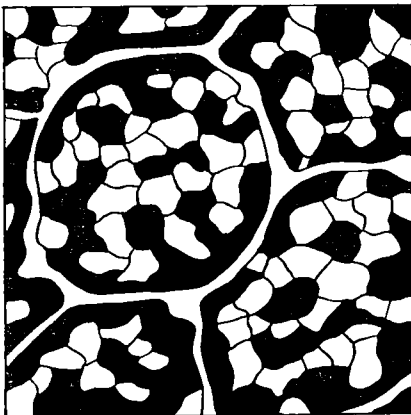


Fig. 73. Microstructure of coarse grain steel which was underheated in annealing, 100X.

If the steel had a coarse-grained structure prior to annealing or normalising, this coarse grain is partly retained after incomplete annealing or incomplete normalising, e. g., in the steel structure shown in Fig. 73, where the original coarse ferrite network has been retained together with the fine grains.

As has been already pointed out, superheating of a steel on annealing and normalising results in rapid austenitic grain growth, particularly in inherently coarse-grain steels, the coarse grains of austenite producing, too, the coarse-grain structure in an annealed and normalised steel (Fig. 74). is characterised by coarse-grain the coarse-grain structure, the

Steel superheated in annealing pearlite and ferrite. Apart from

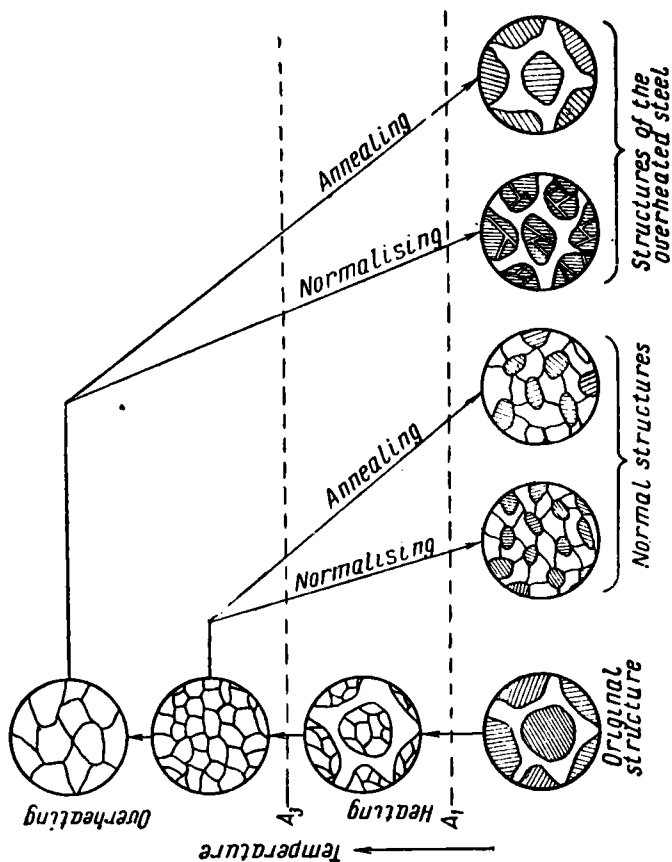


Fig. 74. Diagram showing the change of the steel structure in annealing and normalising.

so-called Widmanstätten pattern (Fig. 75) is peculiar to *superheated normalised steel*. The Widmanstätten pattern is characterised by the ferrite separating not only at the austenitic grain boundaries, but also inside the grains in form of needles. As was shown in an investigation by S. S. Steinberg and W. Y. Zubov, such ferrite is always found in those cases where the ferrite separates from considerably supercooled austenite. The coarse grain produced on superheating and the higher speed of cooling in normalising increase the stability of austenite, which, in turn, is conducive to its supercooling. That is why the Widmanstätten pattern is nearly always found in as-cast steels.



Fig. 75. Widmanstätten pattern of overheated steel, 300X.

Superheating has very little effect on the yield point, tensile strength and hardness of a steel. The per cent elongation and reduction of area somewhat decrease (by 10 to 15 per cent). But the impact strength is severely affected by superheating; its value decreases rapidly as the grain coarsens.

Annealing and normalising operations are not applied as finish treating to improve the mechanical properties of tool

steels, for these steels are used only in as-hardened condition. And annealing is never applied to improve the mechanical properties of alloy structural steels of the pearlitic class. The mechanical properties of these steels are improved through hardening and high tempering, or through normalising (which is rather rare).

31. IMPROVEMENT OF MACHINABILITY

High machinability of a steel is one of the conditions ensuring high productivity in machining. Three main factors of the machining operation are as follows:

- 1) cutting speed;
- 2) cutting force;
- 3) finish of the machined surface.

The faster the cutting speed allowed in machining, the lower the cutting force required and the better the quality of finished surface resulting, the higher is the machinability of the metal.

The machinability of a steel depends upon its mechanical properties. Hard and especially very hard steels are machined poorly, since high cutting forces are required for tools to be cut in the steel being machined. The cutting faces of a tool exposed to the high forces of cut are rapidly blunted, thereby increasing still more the cutting force, and the cutting faces become notched and dull. To prevent a drastic drop in the life of a tool the cutting speed has to be reduced.

Soft steels that usually possess high ductility and toughness are also difficult to machine, though for another reason. When ductile and tough steels are machined a long continuous chip in the form of a helix is produced, which, being in frictional contact with the tool cutting face, brings about its heating, so that the tool becomes worn. The life of the tool is therefore drastically reduced. The machined surface itself is rough, uneven and torn, i. e., its quality is poor.

Numerous investigations have shown that the machinability of low-carbon steels (grades 10, 15, 20, etc., up to 40), including low-carbon alloy steels of the pearlitic class (grades 15Г, 15Х, 20ХГ, 15ХФ, 12ХН2, and others), is at its best after normalising at elevated temperatures (about 900°C). Normalising at elevated temperatures is intended to produce the coarse-grain structure, which is conducive to some decrease in impact strength and a slight increase in hardness which, in turn, improves the conditions of chip breaking.

The highest possible machinability of medium-carbon steels (0.4 to 0.7 per cent of carbon), including the alloy steels (grades 45, 50, 50Г, 40Х, 40ХН, 35ХМА, etc.) is obtained through annealing, which transforms pearlite into a distinctive (differentiated) lamellar structure with ferrite grains located between the grains of pearlite in the form of a broken network.

It has already been shown that it takes a very long time to anneal medium-carbon steels. Therefore the machinability of these steels (e. g., grades 40ХГА, 40ХФА, 12Х2Н3МА, 30ХГС, 38ХМЮА, etc.) is improved through annealing with subsequent high tempering at temperatures between 650 to 680°C. In high tempering the coagulation and spheroidising processes take place in a steel, the structure of which is more distinctive (differentiated) as a result.

Finally, the machinability of high-carbon tool steels is at its best when the structure is composed of grained pearlite (Fig. 76). All alloy tool steels, including those of the carbide class, as well as ball bearing steels (grade ШХ15, etc.), should have a structure of globular pearlite in the as-deliverable state.

The structure of globular pearlite is obtained by means of a special type of annealing. For the globular pearlite to be produced the steel is heated to a temperature of only 20 to 30°C above the A_1 point (carbon tool steels, for instance, are heated to 740 to 760°C), and then very slowly cooled in a furnace at a rate of not more than 50°C per

hour to a temperature of 550 to 600°C. The subsequent cooling may be conducted also in still air at room temperature. The main condition for obtaining the globular pearlite is the accurate maintenance of the given heating temperature. With a considerable excess of the annealing temperature at best mixed (globular and lamellar) pearlite will be formed, and at the worst the result will be purely lamellar pearlite, however slow may be the subsequent cooling. As shown in Fig. 77, the annealing temperature determines the values of the mechanical properties, too. The maximum ductility (elongation and

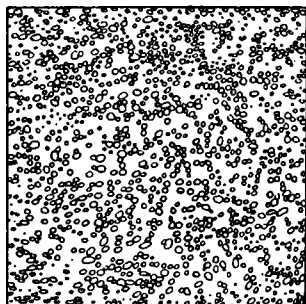


Fig. 76. Globular pearlite, 500 $\times$.

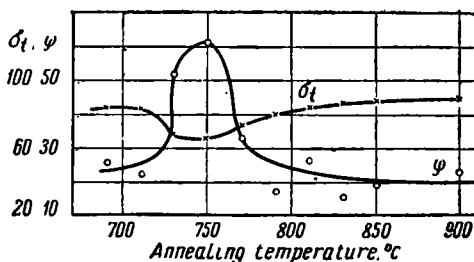


Fig. 77. Diagram showing relation of mechanical properties of steel of grade V8 to the annealing temperature.

reduction of area) and the minimum value of tensile strength in a tool steel of grade V8 are obtained as a result of annealing at a temperature of 750°C, i. e., at exactly the temperature ensuring the formation of the globular pearlite structure in annealing.

A slow cooling in annealing for the formation of globular pearlite is necessary in order to prevent the considerable supercooling of austenite and to ensure its decomposition at the highest temperatures possible below the A_1 point. And the higher the temperatures, the more violent are the coagulation and spheroidising processes.

For the formation of globular pearlite the so-called *pendulum annealing* is sometimes used, with the following sequence of operations: a steel is heated to 750° C, held at that temperature for a short period of time, cooled in furnace to a temperature of 680 to 700°C, then again heated to 750°C and again cooled, etc., several times in succession. During the heating cycles only the small grains of cementite dissolve in the austenite, but there is insufficient time for the larger cementite grains to dissolve. In subsequent cooling cycles the molecules of cementite are deposited mainly on the cementite grains that

are not dissolved in the austenite: hence the coagulation process occurs.

Pendulum annealing is more difficult to perform than an ordinary annealing operation for the formation of globular pearlite, but it takes less time.

32. INCREASE IN DUCTILITY OF STEEL

When a steel is cold-worked (the term "cold-worked" refers to steel that is cold-drawn or cold-rolled), the so-called cold-hardening takes place, i. e., the steel becomes stiffer, its elongation and reduction in area characteristic of ductility markedly decrease, while the hardness and particularly the elastic properties considerably increase.

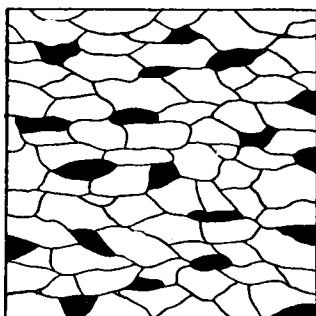


Fig. 78. Microstructure of the steel in cold-hardened condition, 300 $\times$.

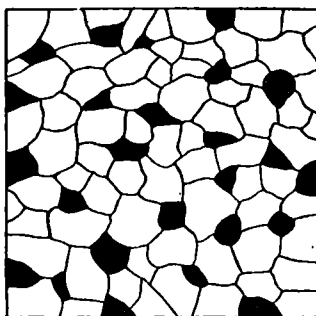


Fig. 79. Microstructure of the cold-hardened steel after recrystallisation annealing, 300 $\times$.

The change in properties under cold working is explained by changes in the steel structure. First, at low reductions (in cold drawing or cold rolling), only the distortion of the space lattice occurs. At greater degrees of deformation, one part of a grain is displaced in relation to another. This displacement gradually involves an increasingly greater number of grains, which begin to "break down" or "refine", progressively elongating in the direction of the strain.

The ductility of a steel may be restored and cold hardening removed through an ordinary phase annealing operation, but that of a cold-worked steel is more often restored by means of a special *recrystallisation annealing*. In this operation new equiaxed grains gradually grow from the fragments of the original elongated grains, forming centres, or nuclei. These new grains progressively become equiaxed (Fig. 78 and 79).

As we know from the metallography course, the recrystallisation temperatures for pure metals can be computed by A. A. Bochvar's formula:

$$T_{\text{recrystallisation}} = 0.4T_{\text{melting}} \quad (14)$$

($T_{\text{recrystallisation}}$ and T_{melting} are in the absolute system or in Kelvin scale.)

The recrystallisation temperature for pure iron is 450° C.

This temperature may be also considered as the recrystallisation temperature for the low-carbon steels. At this temperature the growth of new equiaxial grains progresses very slowly. Therefore, in practice the recrystallisation annealing is conducted at a temperature of 650 to 680°C.

In recrystallisation annealing of steel very little scaling or decarburisation of the metal occurs. This is one of the substantial advantages of recrystallisation annealing over ordinary annealing. In some cases, however, undesirable phenomena take place in recrystallisation annealing. If a steel has been subjected to cold working at small reductions, e. g., 5 to 15 per cent (as is the case in cold rolling, for example), then a very non-uniform grain growth takes place in recrystallisation annealing, with some grains becoming extremely coarse and sometimes reaching a gigantic size (Fig. 80). The large-size grains produced in the steel structure are apt not to restore its ductility but to decrease it further.

To prevent a marked grain growth in recrystallisation annealing it is necessary either to effect the preliminary cold working with reductions in excess of the critical ones (5 to 15 per cent) or, if this is impracticable, to replace the recrystallisation annealing by an ordinary annealing at a temperature well above the A_1 point.

The necessity of increasing the ductility of spring steel sometimes arises in the process of cold winding or coiling of springs. Hot-rolled spring steel which has not been subjected to special heat-treatment at the metallurgical plant possesses low ductility. This can be increased by high tempering, normalising or annealing to obtain the globular pearlite.

The mechanical properties of highly tempered, spring steel of grade 60C2, 10 mm in diameter are as follows:

	First batch		Second batch	
	Tensile strength	Reduction of area	Tensile strength	Reduction of area
Initial condition	102	0	96	0
Tempering at 500°C . . .	104	0	103	0
Tempering at 600°C . . .	102	19	101	36
Tempering at 700°C . . .	84	46	79	46

But the ductility of a spring steel can be improved by high tempering only when the initial steel was not overheated but only "underquenched" as a result of rapid cooling after hot rolling. But if the low ductility was caused by overheating, the high tempering cannot give positive results. In such a case annealing or normalising has to be performed. The diagram in Fig. 81 shows the relationship between the change of the reduction of area for 10-mm spring steel of grade 60C2 and the annealing and normalising temperatures. From this it follows that the ductility of the spring steel is increased on normalising.

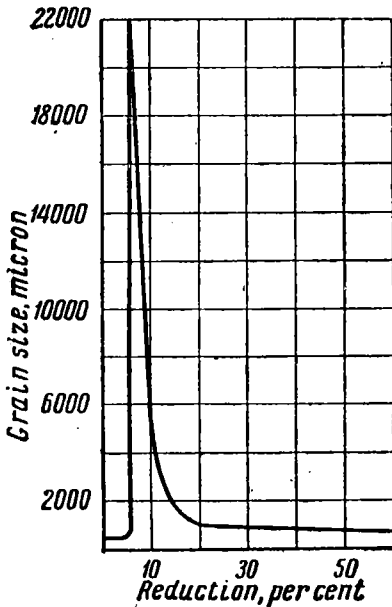


Fig. 80. Diagram showing the effect of reduction upon the grain size of iron subjected to the recrystallisation annealing.

If annealing is used, the annealing temperature should be near to A_1 point (740 to 760°C), i. e., it should be as high as that at which the annealing operation for the globular pearlite was conducted (see Fig. 77).

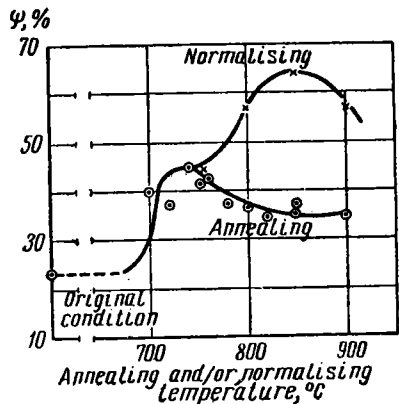


Fig. 81. Diagram showing the effect of the annealing and/or normalising temperature upon the ductility of a spring steel of grade 60C2.

It should be noted that to ensure satisfactory conditions for spring winding the reduction of area must exceed 40 per cent. If the reduction of area for spring steel is lower than 30 per cent, then long practice has shown that the spring winding operation becomes almost impracticable. With a reduction of area ranging from 30 to 40 per cent the winding or coiling of a spring is often accompanied by snapping.

Steel subjected to cold drawing operations is also annealed so as to obtain the globular pearlite structure to increase its ductility.

33. REMOVING OF CHEMICAL NON-UNIFORMITY (DIFFUSION ANNEALING — HOMOGENEISATION)

In some cases the chemical composition of the austenitic grains is non-uniform. This is called intracrystalline liquation or segregation, and is caused by a peculiar feature of the primary crystallisation of solid solutions. As is known from the theory of crystallisation of solid solutions (see the *Metallography Course*), the central portions of the solid solution grains formed at the beginning of crystallisation contain less carbon than the peripheral portions. Similarly, the content of alloying elements in different portions of the austenitic grains in alloy steels is different.

The non-uniformity of chemical composition in austenitic grains is especially marked in ingots and castings. In the course of forging and rolling ingots this chemical non-uniformity is usually removed because at high temperatures diffusion of carbon and alloying elements takes place in the grains of austenite, the chemical composition of which gradually equalises. The plastic deformation facilitates the diffusion processes, because the austenitic grains are broken down thereby, and their different portions move in relation to each other. Therefore, as a rule, in forgings and rolled stock the chemical non-uniformity is not retained. But there are exceptions. Sometimes, a markedly developed lamellar structure can be observed on longitudinal sections cut from the rolled stock (Fig. 82). This is observed especially when the steel is contaminated with slag. On longitudinal sections cut from alloy tool steels in as-rolled condition even the carbide lamellar structure may be sometimes observed (Fig. 83). The photomicrograph shown in Fig. 83 refers to a steel of grade XГ, i. e., to a steel of the pearlitic class. In the steel of this grade there would be no carbide non-uniformity such as is often observed in steels of carbide class (high-speed steels) if the austenite were homogeneous; at a temperature above the A_1 point the structure of steels of the pearlitic class is composed of austenitic grains only. It is evident that the carbide lamellar structure in tool steels of the pearlitic class results from the chemical non-uniformity of the austenite.

The heterogeneity of the steel structure is undoubtedly an undesirable phenomenon, since it gives rise not only to non-uniformity of mechanical properties but to the development of different hardenability in adjacent portions which are different in regard to structure. In particular parts made of steels, the structure of which is represented in Fig. 82 and 83, have produced cracks in quenching.

Chemical non-uniformity can be removed by means of a heat-treating operation termed *homogenising* (equalising) annealing or homogeneisation.

Ingots of alloy steels are usually homogenised at a temperature of 1100 to 1200°C for 10 to 20 hours with subsequent slow cooling.

Steel castings (mainly of alloy steels) are subjected to the same homogenising annealing. Homogenising naturally causes a very rapid growth of austenitic grains. After homogenising, therefore, the steel ingots necessarily undergo an ordinary phase annealing for the fine-grain structure to be formed.



Fig. 82. Microstructure of a rod 100 mm in diameter made of steel of grade 45. The longitudinal section is shown, 100 $\times$.

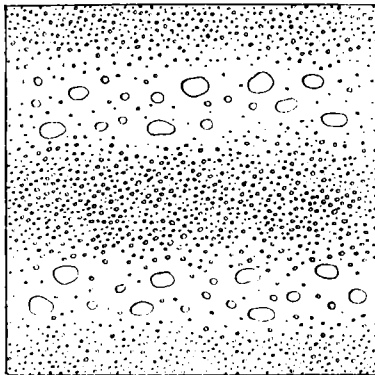


Fig. 83. Microstructure of a rod made of steel of grade XГ. The longitudinal section is shown, 500 $\times$.

The lamellar structure in rolled *steels of pearlitic class* can usually be removed through annealing at a temperature of about 1000° C for several hours; for hypoeutectoid carbon steels a holding time of heating of 1 to 2 hours is sufficient, while for hypereutectoid alloy steels a period of 5 to 6 hours is necessary. It is obvious that to remove the effects of overheating and to restore the steel to its normal condition ordinary annealing or normalising (for hypoeutectoid steels) and normalising with subsequent high tempering (for tool steels) have to be applied.

However, it seems impossible to remove the carbide lamellar structure in steels of the carbide class, e. g., high-speed steels, by means of heat-treating.

34. PREPARING THE STRUCTURE OF STEEL FOR HARDENING

Hardening is the most complicated heat-treating operation. An inaccurate control of the specified hardening temperature generally causes defects, of which the most undesirable are the hardnesses

deviating from the specified values and the formation of quenching cracks.

The principal requirement that should be met by the structure of the steel to be hardened is that it be homogeneous and fine grained. If the original steel structure is coarse grained, then in hardening there will be, as a rule, non-uniformity of hardness, a greater degree of distortion, and even quenching cracks. In the previous paragraph two examples were given where the quenching cracks were said to have been caused by the unsatisfactory original structure of a steel. Prior to hardening the steel must have the fine-grained structure.

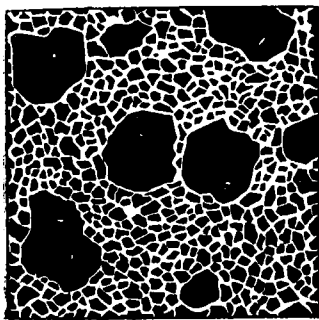


Fig. 84. The non-uniform structure of steel of grade 45, which had very coarse grain and was subjected to normalising at the temperature of 870°C for 1 hour, $100\times$.

This prerequisite is sometimes puzzling. What can be the significance of the size of the initial steel grain, when in heating above the A_1 point the resulting austenitic grains are always found to be fine (in hypoeutectoid steels), whatever the structure of the original steel may be? The austenitic grains are always fine, but they will have a different chemical composition if formed from coarse grains of pearlite and ferrite. The austenitic grains obtained will be non-uniform in chemical composition, some containing less carbon

than others. It is natural that the subsequent transformation processes on quenching, too, will vary.

Fig. 84 is a good illustration of the non-uniformity of carbon content in austenite. It shows the microstructure of a steel with a very coarse original grain. After normalising with a short period of holding, a very heterogeneous structure has resulted; in some areas it has a carbon content of 0.4 to 0.5 per cent, in others 0.8 per cent. Obviously there is a similar distribution of carbon in austenite, too.

Hence parts made from a steel with original coarse-grain structure must be subjected to normalising with a sufficient holding time of heating prior to hardening. The normalising refines the grain structure of a steel; in heating for hardening the grains of austenite become uniform in chemical composition, and none of the previously mentioned defects will occur on hardening.

The original structure of tool steels to be hardened should also conform to special requirements. To ensure the development of a high and uniform hardness in as-hardened condition the structure of steel must be not globular pearlite, but lamellar pearlite. This is

explained as follows: the transformation of globular pearlite into lamellar pearlite (Fig. 13), but the long holding of tools at hardening temperatures that would be sufficient to dissolve all the nuclei of cementite in the austenite is undesirable because, apart from all other drawbacks, it may give rise to oxidation and decarburisation.

Besides, a steel with a globular pearlite structure has inferior hardenability; the cementite grains remaining undissolved accelerate the subsequent decomposition of the austenite. In some cases, where only the surface hardening is required, this is considered to be an advantage, e. g., in the hardening of screw taps. In this operation the exterior portion of a screw tap is hardened, whereas its core remains non-hardened, as is specified.

But if it is necessary to harden a tool to a greater depth or even throughout, then the globular pearlite proves to be undesirable and the pearlite must be changed from the globular condition to the lamellar one before hardening. This aim is achieved through the normalising operation which is generally adopted in the heat-treating of drills: they undergo normalising prior to hardening.

Normalising to obtain the lamellar pearlite is carried out at a temperature of 780 to 800°C.

Hypereutectoid tool steel has to be prepared for subsequent hardening also in cases where the cementite network has been formed or retained for some reason in its structure (Fig. 85). The cementite network sometimes exists in rolled stock, but more often it develops as a result of a faulty forging operation. Most frequently the cementite network is produced in carburising the steel.

The cementite network present in the structure of the hardened tool steel constitutes a very great danger, because in such a form it envelops the steel grains, thereby further increasing the brittleness of the tool steel, which is already fairly brittle. The result is as if each grain of the steel structure were surrounded by cementite in the form of a brittle envelope. It is necessary therefore to eliminate this cementite network in the steel structure, and if it has developed for

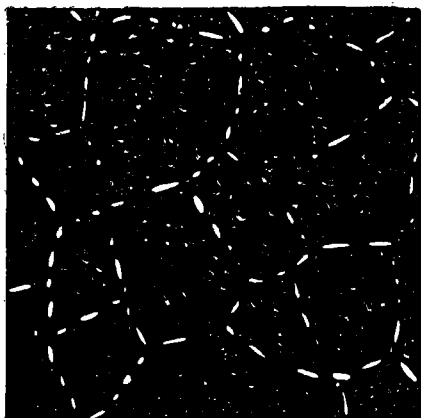


Fig. 85. The cementite network in hardened hypereutectoid steel. 500X.

some reason or other, it should be removed. This objective can be also attained through the normalising operation. During relatively rapid cooling in still air the cementite network has insufficient time to develop, and the cementite separates in the form of distinct grains.

The normalising temperature to eliminate the cementite network in hypereutectoid tool steels should exceed the A_1 (A_{cm}) temperature and be equal to:

Grade of the steel	Normalising temperature
Y9, Y9A	800° C
Y10, Y10A	850° C
Y11, Y11A	875° C
Y12, Y12A	900° C
Y13, Y13A	950° C

35. RELEASE OF INTERNAL STRESSES

Ordinary, or phase, annealing is used to relieve internal stresses in steel castings and forgings. Those in welded structures cannot be removed through an ordinary (phase) annealing. This is because in an ordinary annealing operation, i. e., on heating at temperatures of about 850°C, the elastic properties of a steel are decreased in such a degree (see page 85) that the welded structure may distort under the load produced by its own weight. For this reason welded structures are generally not subjected to ordinary annealing.

To relieve internal stresses in welded structures a special annealing operation is applied, consisting of heating at temperatures sufficient for local plastic deformations to occur under the effect of internal stresses which, as we have seen (Chapter V), takes place at temperatures from 500 to 600°C.

Annealing of welded structures for releasing of internal stresses is conducted as follows:

1) slow heating at a rate of 100 to 150° per hour up to a temperature of 600 to 650°C;

2) holding at that temperature for a period of time determined on the basis of the following calculation: a period of 3 to 4 minutes for each millimetre of the maximum thickness of steel sheets the welded structure is composed of;

3) cooling at the rate of 50 to 100° per hour to a temperature of at least 300°C. The slow cooling is necessary to prevent the setting up of new internal stresses on cooling, which inevitably develop if the steel is cooled rapidly because this always progresses non-uniformly. There are many cases recorded where the cracks have developed in annealed welded structures simply because they have been non-uniformly cooled.

36. ISOTHERMAL ANNEALING

Knowledge of the kinetics of supercooled austenite transformations at temperatures below the A_1 point makes it possible to perform an accelerated annealing operation at a constant temperature, the so-called *isothermal annealing*. Fig. 86 gives a diagram illustrating the transformation of supercooled austenite for chromium ball bearing

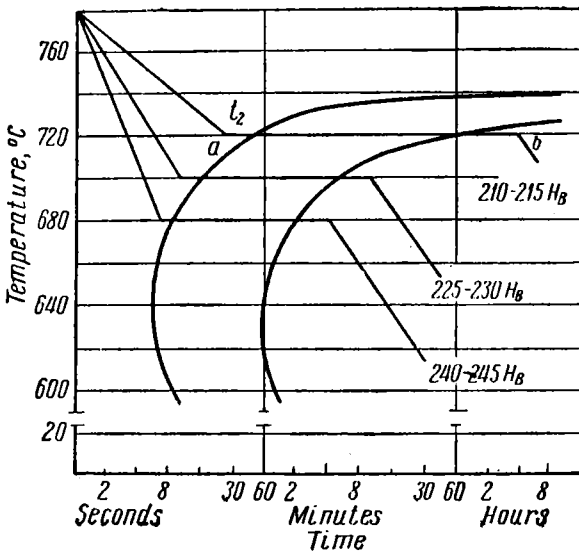


Fig. 86. Diagram showing the isothermal transformation of supercooled austenite in steel of grade IX15 (Y. R. Raizin).

steel of grade IX15. The steel is heated to a temperature t_1 (above the A_1 point), then cooled comparatively rapidly to a temperature t_2 , and finally held at that temperature for a period of time determined by a section ab on the diagram. In this process the austenite in the steel decomposes fully into a ferrite-carbide aggregate. From the above diagram it is clear that the lower the isothermal annealing temperature (of course, to the inflection point of curves), the shorter will be the duration of the isothermal annealing. This is a very tempting advantage of isothermal annealing over ordinary annealing, since instead of being annealed for several hours a steel can be treated for a much shorter time, e. g., for 1 to 2 hours (the heating time

not being counted). The fact that after the isothermal annealing the steel may be cooled rapidly, in air for instance, is also an important advantage of this heat-treating operation.

Unfortunately, the above-mentioned advantages of the isothermal annealing are to some degree lost because of the fact that the lower the transformation temperature (to a temperature corresponding to the inflection point of curves) and the shorter, consequently, the duration of the operation, the higher is the hardness of the resultant ferrito-carbide conglomerate. But if the temperature of the isothermal annealing be increased to a level at which a minimum hardness of the ferrito-carbide conglomerate results then, as is evident from the diagram in Fig. 86, the duration of the isothermal transformation will markedly increase and the isothermal annealing will have no advantages whatsoever over ordinary annealing. In cases, however, where the specifications of the hardness of annealed steel are less exacting and higher values of hardness are permitted, the transformation temperature of the isothermal annealing may be lowered. The duration of the isothermal annealing operation will then be greatly shortened, becoming less than the time required for carrying out ordinary annealing. In those cases the isothermal annealing is to be applied.

Chapter VII

HARDENING

37. PURPOSES OF HARDENING

The operation of hardening is applied to all tools and some important machine parts intended for especially heavy-duty service, as well as to all machine parts made of alloy steels.

The main purpose of hardening tools is to develop *high hardness*. The main object of hardening machine parts made of structural steels of the pearlitic class is to *increase* their tensile strength and, in particular, *elasticity* (which is characterised by the *yield point*). As we have seen in the course on the *Strength of Materials*, the especial significance of the yield point increase lies in the fact that it determines the values of maximum allowable stresses, for in general the higher the yield point, the higher are the allowable stresses and, consequently, the higher working stresses may be considered safe to apply to machine parts when in service. Thus, the *increasing of the yield point* characteristic of the steel elastic properties constitutes one of the main objectives of heat-treating, particularly of the hardening operation.

But, as is known from Fig. 46, the high values of tensile strength and yield point are brought about not in hardened condition, when the steel structure is composed of tetragonal martensite, but after tempering, when the steel structure becomes two-phase with a high degree of dispersion. Therefore, strictly speaking, the hardening operation as applied to structural steels is aimed only at preparing the structure for those transformations which take place on tempering. The high values of tensile strength and yield point are peculiar characteristics of the tempered structures obtained in the decomposition of the hardened martensite. It should also be noted that in tempering there occurs the increase in ductility and toughness of a steel, the value of which in as-hardened condition is nearly zero.

So high mechanical properties (*strength, elasticity, ductility and toughness*) in articles of structural steels can be brought about through a complicated heat-treating operation consisting of *hardening*, followed by *high tempering*. Such a heat-treating operation has been termed *hardening with subsequent tempering*, as was previously pointed out.

By hardening parts of structural steels their hardness is also increased. In some cases this fact is of no significance at all. In other cases the increase in hardness is detrimental because the machining of parts is made very much more complicated thereby, even after tempering.

However, in several cases the development of high hardness becomes the main object of the hardening of machine parts. For example, those parts that are subjected to wear in service, such as shafts, gears, etc., need high hardness. Under otherwise equal conditions, the higher the hardness, the higher will be the *wear resistance* of the metal. The hardening of articles increases their surface hardness and therefore brings about an increase in the wearing resistance of these parts. The increase in surface hardness is accomplished through a special surface hardening operation or quenching after preliminary chemical-thermal treatment.

38. PREPARING ARTICLES FOR HARDENING

In some cases the heating of parts for hardening should be preceded by a certain preparatory operation.

Foreign matters present on the surface of articles and tools (impurities, scale, etc.) may drastically decrease the hardening effect, particularly in those cases where the high surface hardness has to be obtained. In such cases it is necessary to clean the surface before they are heated. Oil or grease impurities are removed by rinsing in the hot water (preferably with soda added), and the pieces of scale

or other foreign substances are cleaned off with wire brushes or in a sand-blasting machine.

Holes in articles and tools to be quenched in water may prove to be causes of cracks. These holes are therefore stopped with wet asbestos mass. If there is a thread in these holes, they may be closed by screwing plugs in them.

It is necessary to provide for the manner in which the heated parts or tools are to be immersed in the quenching tank. Small parts are

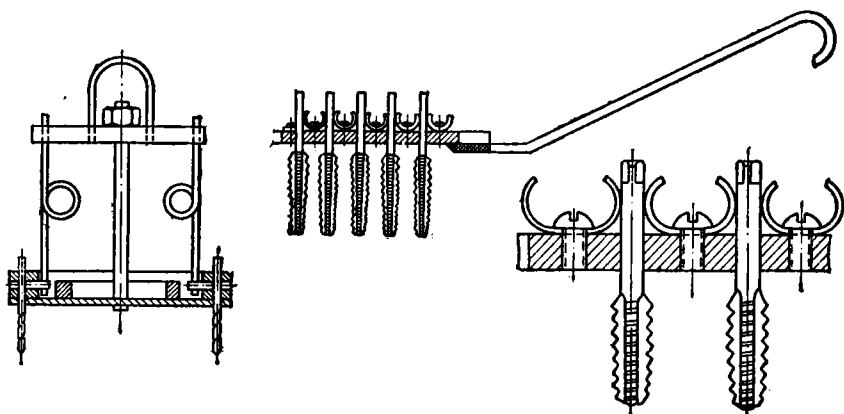


Fig. 87. Multi-seated fixtures for quenching small-size, slim tools (bits, screw taps, etc.).

often heated in pans, on iron sheets, etc., and are simply poured into the quenching tank, in which the netted basket is immersed. For some small-size, thin articles and tools having relatively great length as compared with their cross-section (small bits, screw taps, etc.), such a method is, obviously, unsuitable, as distortions may occur in pouring and non-uniform hardness may develop. These parts must be immersed in the quenching tank in an exactly vertical position. To ensure such a manner of immersing the above-mentioned parts various multi-seated quenching devices are employed (Fig. 87).

Articles or tools of larger size to be subjected to local hardening are withdrawn from the heating furnace by the operator, who takes them with tongs by the end (Fig. 88) and thus transfers into the quenching tank. However, this method is unsuitable for parts and tools that must be hardened throughout, because soft spots are formed on those areas which are in contact with the cheeks of the tongs. At the worst the great difference in cooling rates can even give rise to the formation of cracks on these spots. Special tongs with sharp bits (Fig. 88, *d*) or centre punches are employed in quenching such articles. If the parts or tools to be quenched have holes, they can be

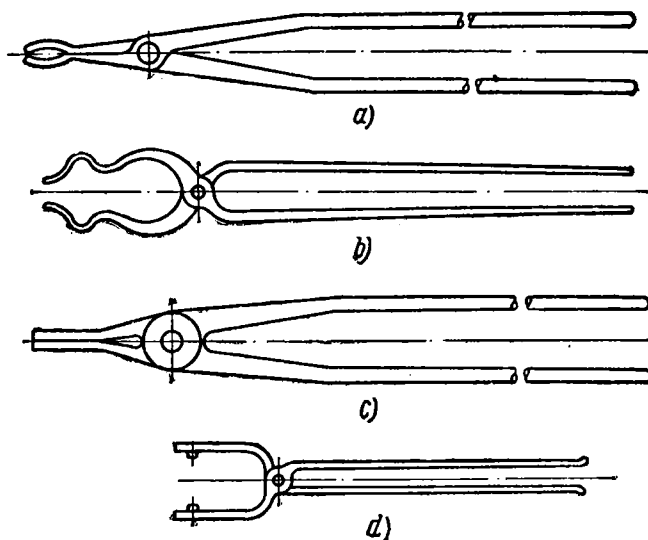


Fig. 88. Hardening tools.

a and b—for articles and tools of cylindrical form; c—for flat articles and tools; d—for flat articles and tools of which the whole surface must be hardened.

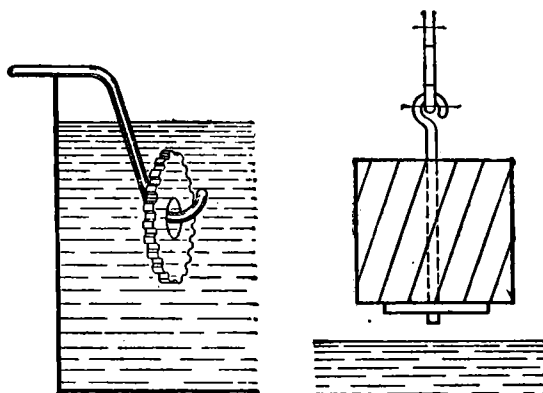


Fig. 89. Quenching fixtures for milling cutters and other tools having holes.

tied round with wire. In such cases special hooks or suspensions may also be used to immerse the articles (Fig. 89).

To prevent distortion of springs of great length and comparatively small diameter when heated and cooled in hardening they are tightly fitted on hollow mandrels, which are sections of thin-walled pipes. Helical extension springs closely coiled (without a space between coils) should be drawn together with wire along the axis in order to avoid their "parting" (pulling apart) on heating.

It is evident that in preparing articles for hardening they must be properly protected from oxidation and decarburisation (see Chapter V).

39. HARDENING TEMPERATURE FOR THE PEARLITIC CLASS STEELS

In hardening any steel, both hypoeutectoid or hypereutectoid, it should be in each case heated to a temperature above the A_1 point, i. e., to a temperature at which austenitic grains instead of pearlitic ones are found in the steel structure.

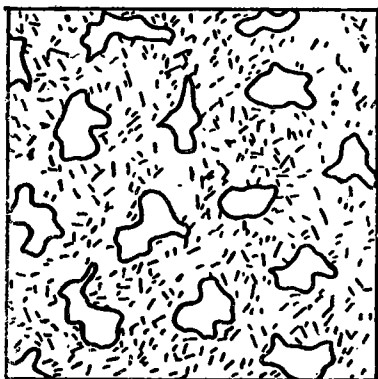


Fig. 90. Microstructure of hypoeutectoid steel subjected to an incomplete hardening. Martensite and ferrite, 500 $\times$.

The structure of each hypoeutectoid steel, when heated above the A_1 temperature but below the A_c point, is to be composed of austenitic and ferritic grains. If a steel heated to such a temperature is quenched in water the martensitic transformation will take place in the austenitic grains. No transformations will occur in the grains of ferrite during quenching. As a result of incomplete hardening of hypoeutectoid steel by quenching from a temperature of between A_1 to A_c , the steel structure will consist of martensite and ferrite (Fig. 90). Such a structure will prove unsatisfactory in all respects. In fact, if the

hardening operation is aimed at obtaining high hardness, then the grains of soft ferrite present in the steel structure will not permit the attainment of this aim, and the hardness of the steel will be decreased. If the purpose of the hardening operation is to increase the strength and elasticity of a steel, then again the presence in the structure of grains of ferrite having low strength and elastic properties will decrease these characteristics in tempering, thus preventing the development of high tensile strength and elasticity.

For these reasons the incomplete hardening of hypoeutectoid steels is not used in practice.

Hypoeutectoid steels are subjected in general only to full hardening, i. e., they are quenched from temperatures above the A_3 point.

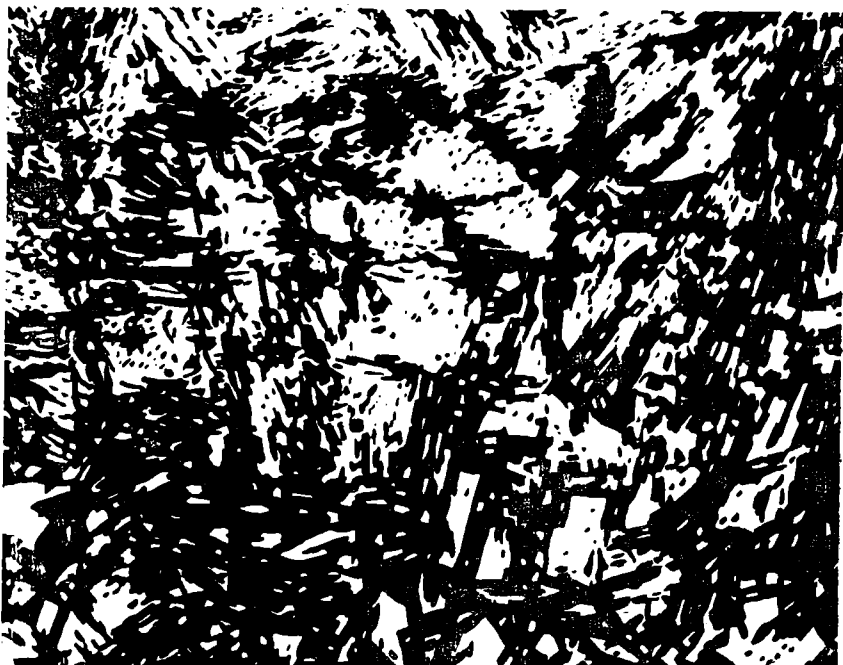


Fig. 91. Microstructure of steel overheated in hardening. Coarsely acicular martensite, 500 $\times$.

At such temperatures the steel structure is entirely composed of austenitic grains. Therefore the structure of a steel quenched in water from a temperature above the A_3 point consists of martensite.

The optimum hardening temperature to be accepted for treatment of hypoeutectoid steels may be determined from the following equation:

$$t_{\text{quench}} = A_3 + (30-50^\circ\text{C}). \quad (15)$$

Quenching of hypoeutectoid steels from temperatures considerably above the mentioned optimum leads to their overheating. Overheating in hardening manifests itself in the martensite's resulting in the coarsely acicular form (compare Fig. 91 with Fig. 24). Steel

overheated in hardening, which has a structure of coarsely acicular martensite, has an extremely high brittleness.

Unlike the hypoeutectoid steels, the hypereutectoid ones are subjected, as a rule, only to incomplete hardening. The optimum temperature to be accepted for hardening hypereutectoid steels is calculated with the following formula:

$$t_{\text{quench}} = A_1 + (30-50^\circ\text{C}). \quad (16)$$

At this temperature the structure of any hypereutectoid steel consists of austenitic grains and small particles of secondary carbides. On quenching from such temperatures the structure of the hypereutectoid steel is composed of martensite and the secondary carbides (Fig. 92). Since the hardness of carbides (about 800 H_B) is greater than that of martensite (650 to 750 H_B), and since tool steels are hardened in order to obtain a maximum hardness, the operation of in-

Table 14

**The Effect of the Hardening Temperature
upon the Hardness and Acicularity of the Martensite
in Grade Y12 Steel Being Quenched in Water**
(100 specimens were tested at each hardening temperature)

Hardening temperature, °C	Rockwell C hardness number	Length of the martensitic needles, mm	Number of specimens with quenching cracks
740	65	Not determinable	0
760	65	" "	0
780	65	" "	30
800	64	Less than 0.005	50
820	63	0.005 to 0.01	80
840	62	0.01 to 0.02	100
860	62	0.01 to 0.02	100
880	61	0.03 to 0.04	100
900	60	0.04 to 0.06	100

complete hardening is suitable for increasing the hardness of hyper-eutectoid tool steels to the maximum.

The heating of hypereutectoid steels to higher temperatures for hardening causes a gradual dissolution of carbides. On heating to a temperature above the A_1 (A_{cm}) point the steel structure consists only of austenitic grains. These grow larger and, consequently, the martensite becomes more coarsely acicular (Fig. 93). The hardness of hypereutectoid steel in as-hardened condition somewhat decreases as the hardening temperature rises; this is explained by the fact that, firstly, there are no more solid carbides present in the steel structure, and, secondly, the amount of residual austenite grows larger as the hardening temperature is increased.

The effect of the hardening temperature upon the structure and hardness of the grade Y12 steel is shown in Table 14.

It is evident from Fig. 93 and Table 14 that carbon tool steels are very sensitive to overheating. It should be noted that one of the advantages of alloy tool steels (in particular grades XΓ and XBr) is that the martensite retains the

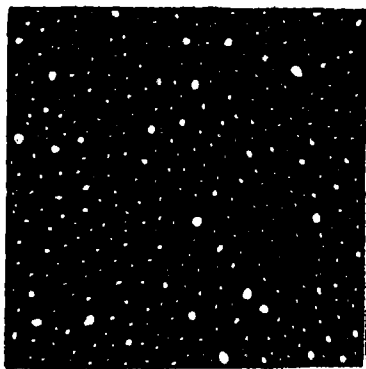


Fig. 92. Microstructure of normally hardened and tempered hypoeutectoid steel. Tempering martensite (with no signs of acicularity) and carbides, 500 $\times$.

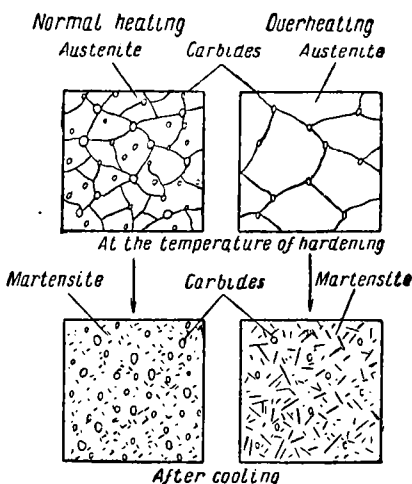


Fig. 93. Effect of the hardening temperature for hypereutectoid carbon steel upon the amounts of carbides and acicular martensite in the steel structure. On overheating martensite is coarsely acicular, while the amount of carbides considerably decreases.

finely acicular structure when somewhat overheated in hardening, while the hardness nearly does not decrease with the increasing of temperature.

Table 15 shows the hardening temperatures for the most important carbon and alloy steels of the pearlitic class.

The hardening temperatures for steels of the austenitic and carbide classes will be given in one of the following paragraphs.

Table 15

Hardening Temperatures for the Main Pearlitic Class Steels of Standard Grades

Grade of steel	Hardening temperature, °C	Grade of steel	Hardening temperature, °C
<i>Structural steels</i>			
Carbon steels			
Cr. 4; 20; 25	870 to 890	40XH; 45XH;	810 to 830
Cr. 5; 30	850 to 870	50XH	810 to 830
35	840 to 860	20XH3A; 30XH3	870 to 890
40	830 to 850	20X2H4	940 to 960
Cr. 6; 45	820 to 840	18XHBA	840 to 860
Cr. 7; 50; 55	800 to 820	25XHBA	840 to 860
60; 65; 65Г	790 to 810	40XHMA	840 to 860
Alloy steels		<i>Tool steels</i>	
		Carbon steels	
20X; 30X; 35X	850 to 870	Y7	770 to 790
40X	840 to 860	Y8, Y9, Y10, Y11,	760 to 780
45X	830 to 850	Y12	
50X	820 to 840	Alloy steels	
20XΦ; 40XΦA	870 to 890	7X3; 8X3	850 to 880
50XΦA	850 to 870	9X	820 to 850
20XM	870 to 890	9XC	845 to 865
30XM	860 to 880	XГC	820 to 860
35XM	840 to 860	5XBГ	850 to 900
35X2MA	860 to 880	X; X09	830 to 850
40XC	870 to 890	X05	800 to 825
20XГ	850 to 870	Φ	820 to 840
20XГC; 25XГC	870 to 890	B1	810 to 830
30XГC; 35XГC	870 to 890	XГ	830 to 850
18XГM	850 to 870	9XBГ and XBГ	825 to 845
40XГM	840 to 860	XB5	820 to 860
35X10A; 38XM10A	930 to 950		
20XH	830 to 850		

Notes: 1. The hardening temperatures for high-quality steels (with symbol A) are the same as those for the corresponding quality steels (without symbol A).

2. The hardening temperatures for spring steels and steels for hammer dies are given in Chapter XV-77 and Chapter XVI-83, respectively.

40. INTERNAL STRESSES SET UP IN QUENCHING

Cooling in quenching progresses rather *ununiformly*, with the surface of the metal cooling very rapidly and the central portions somewhat slower. The uneven distribution of temperature on the cross-section of a cooled part causes non-uniform volumetric changes, with the surface layers of the article contracting more rapidly than its internal portions. The compression of the outer layer is thus inhibited by the central ones. It therefore follows that a rapid and consequently non-uniform cooling will throw the central portion under *compression* and the outer layers under *tension*. When the surface has fully cooled and its contraction has ceased, the central portion will continue to contract. Thus, the tensile stresses in the outer layers and the compressional ones in the central portion will progressively grow smaller. It may also happen that the stresses in the outer layers will become compressional after reaching zero, whereas those in the central portion will become tensile.

The stresses that develop in a rapidly cooled article as a result of an unequal cooling are the so-called *thermal stresses*. The thermal stresses are developed in any article, irrespective of the material it is made of, whether it be steel or copper.

Apart from thermal stresses so-called *structural stresses* are set up in rapidly cooling parts made of alloys that, like steels, are subject to structural transformations in a solid condition. These structural stresses are caused by two factors:

- 1) the unequal specific volumes of austenite and its decomposition products;
- 2) the structural transformations progressing at different times in the outer layers and the central portion of the article.

The specific volume of the martensite is the maximum, whereas that of the austenite is the minimum, as is plain from the following figures:

	Specific volume, cu cm per gram
Ferrite-cementite conglomerate	0.127
Martensite	0.127 to 0.130
Austenite (at room temperature)	0.122 to 0.125
Austenite (at the martensitic transformation temperature)	0.126 to 0.127

For martensite and austenite the limits of numerical values for specific volume are indicated, because the specific volumes of these structural constituents depend upon their carbon content in the following manner: the higher the carbon percentage, the greater is the specific volume.

Thus, the internal stresses set up in rapidly cooled steel articles being quenched are a combination of thermal and structural stresses.

Now let us consider the diagram illustrating the distribution of internal (thermal and structural) stresses in hardening a steel article throughout. The graphs of Fig. 94 show the volumetric changes in the outer layer (curve S) and in the central portion of the article (curve C) at various moments in the cooling process.

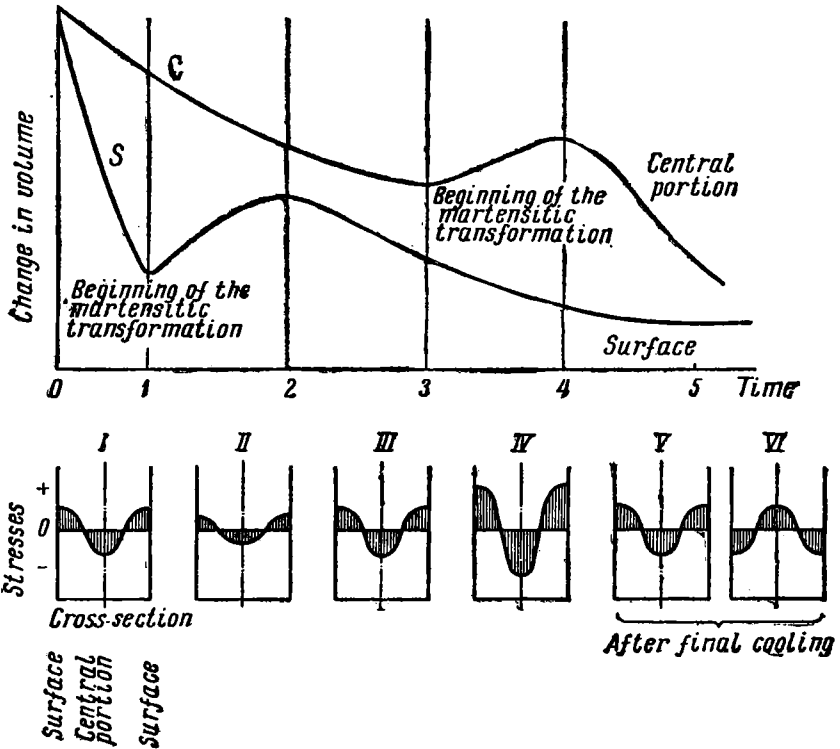


Fig. 94. Diagram showing the volumetric changes and distribution of internal stresses in a steel article being through hardened.

Both the outer layer and the central portion are cooled only in the time interval from 0 to 1. As the temperature drops, the volume contracts. It is natural that the volume of the outer layers contracts at a faster rate than that of the central portion, and this puts the outer layers under tension and the central portion under compression. This is illustrated by graph I.

In the time interval from 1 to 2 the central portion continues to contract while in the outer layers the martensitic transformation takes place and, consequently, some volumetric expansion occurs (it is to be remembered that the specific volume of martensite is greater than that of austenite). As a result of the central portion being under compression and the outer layers under tension, the internal stresses are somewhat lowered (as is shown by graph II). In this stage of the cooling process the structural stresses are opposite to the thermal ones in their effect.

In the time interval from 2 to 3 both outer and central portions are thrown under compression, with internal stresses growing somewhat higher (as is illustrated by graph III). It should be noted here that the martensitic transformation continuously takes place at all temperatures below the M point. It also occurs in the time interval from 2 to 3. This means that in the interval 2 to 3 not only does the volume expand, but, at the same time, thermal compression occurs. As the thermal compression is higher, a contraction of volume results.

In the time interval from 3 to 4 the outer layers continue to be compressed while the central portion of the metal undergoes the martensitic transformation and therefore expands. As a result the stresses are considerably increased (graph IV).

In the time interval from 4 to 5 both outer and central portions contract, while the stresses diminish to some degree (as is shown by graph V) and may even reverse (graph VI).

The diagram showing the distribution of stresses is, naturally, simplified to a considerable extent. Indeed, the graphs indicating the distribution of internal stresses at different moments of the cooling process are more complicated.

From the above diagram very important conclusions may be drawn. The main conclusion is that the internal stresses are highest not after the metal has been completely cooled, but during the cooling process itself, i. e., when the martensitic transformation is taking place in the central portion of the article. Attention should also be paid to the fact that these maximum stresses are *tensional*, i. e., of precisely the only type which is capable of producing cracks (compressional stresses cannot cause the development of cracks).

As we have seen in Chapter V, the internal stresses may be "removed" or removed as a result of local plastic deformations. But it is obvious that plastic deformations can occur only in metal possessing at least some degree of plasticity. Yet, as is evident from the experimental data compiled by A. L. Nemchinsky, N. M. Fokina and I. L. Shimelevich (Table 16), steel containing carbon in excess of 0.4 per cent does not possess plasticity at all, even in the initial stages of the martensitic transformation.

Table 16

**Mechanical Properties of Steels with Various Amounts
of Martensite in Their Structure**

Carbon content, per cent	Amount of martensite in the steel structure, per cent	Tensile strength	Yield point	Reduction of area
0.20 to 0.24	0	77	33	63
	10	117	56	45
	45	167	118	29
	75	175	129	28
	100	194	150	29
0.45	0	81	37	48
	15	75	55	0
	35	74	—	0
	45	66	—	0
	65	43	—	0
	100	52	—	0
0.86	0	69	49	17
	5	60	—	7
	30	63	—	—
	40	63	—	—
	55	42	—	—
	75	22	—	—
	85	19	—	—

The case of hardening throughout has been considered above. When the article has not been hardened throughout, the resultant internal stresses will be somewhat lower because the volumetric changes in its central portion, with austenite transforming into the ferrito-carbide aggregate, are less; in fact the specific volume of the ferrito-carbide aggregate differs from that of the austenite to a lesser degree than does the specific volume of the martensite.

When the internal stresses exceed the yield point of the steel the piece undergoes plastic deformation, i. e., it is distorted or warped. But if the internal stresses exceed the tensile strength of the metal, then cracks will inevitably develop.

The setting up of internal stresses in hardening cannot fully be avoided. They can be only mitigated. For this purpose various cooling methods have been evolved, based either upon the heat-treating practice or the knowledge of the intrinsic nature of the processes that take place in heat-treating operations.

41. METHODS OF COOLING IN HARDENING

Cooling of the article or tool to be hardened is the most difficult and important part of the hardening operation. The difficulties and complexity involved are due to the fact that the problems to be solved by the cooling are of a contradictory nature. On the one hand, in hardening for martensite to be formed the cooling rate employed must be faster than the critical one, and since the critical cooling rate for plain carbon steels is very fast (see Fig. 35), the actual cooling rate for hardening should consequently also be very rapid.

On the other hand, it has been just pointed out that rapid cooling is the main cause of the development of substantial internal stresses, which at best lead to distortions of the articles, and at the worst to the formation of cracks.

These internal stresses are especially dangerous in cooling within the range of martensitic transformation, where they reach their maximum values when the steel practically loses its plasticity completely. From this, obviously, it should be concluded that cooling through the martensitic transformation range must be conducted at the lowest possible rate capable of giving the assigned hardness in the article hardened.

One of the simplest and most commonly used methods of cooling in hardening articles and tools of carbon steels is to quench them successively in two media, first in water and then in oil. This method, which is called briefly quenching "*through water to oil*", consists of first plunging the article or tool into water for a few seconds to remove a part of the heat and then into oil till the cooling is complete.

This old quenching practice can easily be explained on the basis of the known data of the kinetics of transformations that take place in supercooled austenite. What, in fact, is the object of the quick cooling in hardening steel? It is the suppressing of the pearlitic transformation of austenite. But for this aim to be attained there is no need for rapid cooling through the whole range between the hardening temperature and room temperature. It is sufficient to cool the steel rapidly only through the temperature range from A_1 to 400°C , i. e., precisely through that temperature range where the austenite is the least stable and transforms into the ferrite-carbide aggregate. If the decomposition of austenite into ferrite-carbide aggregate is suppressed through a quick cooling, then in further cooling, even though it be comparatively slow, the supercooled austenite can only transform into martensite.

The difficulty of conducting the quenching through water to oil in practice consists in that it is impossible to predetermine by calculation the time a piece to be hardened should be held in water, i. e.,

the precise moment at which the article must be transferred from the water into the oil bath. If the article is held in water for a few seconds too long, the martensitic transformation inevitably occurs. But if, on the contrary, the article is transferred into the oil bath too hastily, when the stability of the austenite is still low, the pearlitic transformation, even though only partial, cannot be avoided, and the part will not develop the desired hardness.

In the case of quenching through water to oil as applied to a batch of articles of the same type, the time of holding in water is ascertained by means of testing a number of samples. But if the individual parts are quenched, then the criterion for determining period of time the article should be held in water is the moment when the tongs used to hold the piece cease to vibrate. The cessation of vibration of the tongs is indicative of the water ceasing to boil violently around the article. The time the piece to be hardened must be kept in water is computed approximately on the basis of 1 second for each 5 to 6 mm of diameter or thickness of the article cross-section.

In hardening parts of heavy sections, the *interrupted quenching technique* is sometimes used, the so-called quenching with "dipping". This method consists of first plunging the article into water (usually by means of some lifting appliance) for a period of several minutes, then withdrawing it to the open air, then again plunging it into water and then again withdrawing it to the open air, repeating this procedure several times. When the piece is withdrawn from water to the open air, it is somewhat heated by the heat conducted from its central portion through the metal to the surface. Because of this the temperature of the part is equalised across its section. Though the internal stresses are really lower in this heat-treating operation, yet the surface hardness obtained is not the highest possible. Therefore this method of quenching is applied only to machine parts or tools, the surface hardness of which is not specified to be very high (e. g., hammer dies).

The internal stresses and strains can be relieved to a considerable degree by means of the broken and isothermal hardening operations. These consist of cooling the article to be hardened not in a cold liquid (water or quenching oils) but in fused salt heated to a temperature slightly exceeding the martensitic transformation temperature.

The broken hardening process is based upon the fact that at temperatures only slightly above the M point the stability of the austenite is high (Fig. 95).

The broken hardening operation consists of three steps or stages: 1) cooling the part in a fused salt bath (line ta), 2) holding it in the fused salt (line ab), and 3) subsequently cooling it in the open air (line bc). The cooling in the fused salt (the ta line) should be conducted at such a rate as to prevent ferrite-carbide transformation. This

means that the cooling line (ta) should be located to the left of the left curve showing the beginning of the austenite transformation. Since the left curve for carbon steels very closely approaches the ordinate axis (see Fig. 17), it is evidently practically impossible to accomplish the broken hardening of articles and tools of carbon steels without some transformation of the austenite into ferrite-carbide aggregate even though this may be only partial. Quite the contrary is true for alloy steels, for which the curve illustrative of the beginning of the austenite transformation is displaced to the right.

When steel is held in the fused salt no transformations occur in it (the ab line). The martensitic transformation begins only after the article has been removed from the fused salt bath and its temperature has dropped to the M point.

To accomplish broken hardening it is necessary that the temperature of the article to be hardened should be equalised across its section before the martensitic transformation begins. The internal stresses developed in the article during its cooling in the fused salt will to a considerable degree be eliminated by local plastic deformation, because of the very high plasticity of the austenite. Thus, at the time the martensitic transformation starts the steel article treated will be almost completely freed from internal stresses and strains. It is true that in the course of the martensitic transformation itself internal stresses will develop but their magnitude will be fairly moderate because the martensitic transformation will proceed nearly simultaneously in both the outer layers and the central portion of the piece being hardened.

Broken hardening is applied to tools of intricate design, with abrupt changes in cross-section, and made of alloy steels, e. g., composite dies and intricate milling cutters of high-speed steels, which develop cracks when quenched in oil.

Isothermal hardening is conducted similarly to broken hardening, the only difference being that in the first operation the part treated is kept in the fused salt bath until the A_r'' -transformation is completed. In the diagram of Fig. 95 the isothermal hardening process is illustrated by the broken line $tabde$.

The steel subjected to isothermal hardening has a structure consisting of acicular troostite. The hardness of a steel isothermally

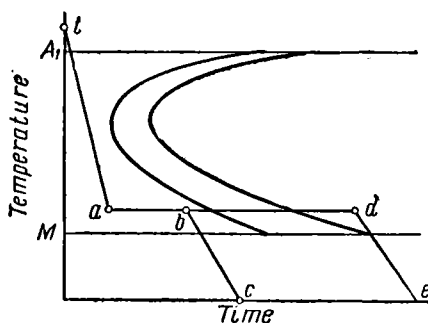


Fig. 95. Diagram showing the broken and isothermal hardening operations.

hardened is lower than that of a steel subjected to the broken hardening operation. The fact that in some cases there is no need to carry out tempering after the isothermal hardening operation constitutes an important advantage of isothermal hardening.

The isothermal hardening process is applied to machine parts.

The new method of hardening, the so-called bright hardening, is of a very great practical interest. In this method the metal is heated in salt baths and subsequently cooled in fused alkalis (e. g., sodium hydroxide, NaOH; potassium hydrate, KOH; and their mixtures). The surface of parts subjected to the bright hardening operation is so clean that after quenching there is no need for any cleaning procedure (e. g., sandblasting); at the same time no diminishing of dimensions as a result of oxidising is noticed. Bright hardening is much more easily performed by heating in salt baths and cooling in fused alkalis than by heating in furnaces with protective gas atmosphere. Therefore the technique of bright hardening in fused alkalis is now extensively used at machine-building plants in the mass production of small parts (e. g., bolts).

According to the temperature of the fused alkali, there are recognised two types of bright hardening—bright isothermal hardening, which is the term applied when the temperature of the alkaline bath is above that of the martensitic transformation process, and bright hot hardening, in which the heating in the alkaline bath is below the martensitic transformation temperature.

The chemical compositions of baths used in the broken, isothermal and bright hardening operations are shown in Table 10.

The hardening operations considered above allow a *general decrease* in internal stresses and strains. *The uniform distribution of the internal stresses* in the whole volume of the article treated is of no less importance. In some cases the non-uniformity of the internal stresses is caused by incorrect design of the part or tool being heat-treated, e. g., by a combination of thick and thin portions, abrupt changes in cross-section, sharp projections, holes of small diameter in the massive parts, etc. All these detrimental factors must be carefully avoided. The more regular and simple the shape of an article or tool, the more uniformly (under otherwise equal conditions) are distributed the internal stresses and strains. But it is not always possible to make the shape of an article or tool simple and regular. For this reason measures should be taken in hardening to bring about as uniform a cooling as possible.

When immersing heated parts or tools in a quenching liquid the following principal regulations should be complied with:

- 1) articles composed of heavy and thin sections must be immersed in the quenching bath with their heavy part first;

2) long, slender articles such as bits, screw taps, reamers and springs must be immersed strictly vertically (Fig. 96) or they will warp;

3) thin, flat parts such as discs and disc milling cutters must be always immersed in the quenching bath edgewise;

4) parts in the form of thin rings should be immersed with their axis perpendicular to the surface of the quenching liquid;

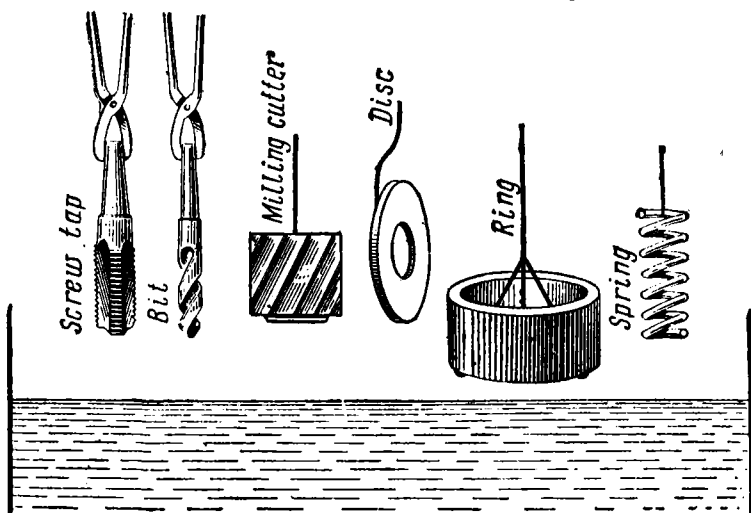


Fig. 96. Sketches illustrating the proper methods of immersing articles and tools to be hardened in the quenching tank.

5) articles with concave surface should not be immersed in the quenching bath with this surface downwards, or a vapour coating will form and prevent the hardening of this part of the surface of the article.

Holes of small diameter in massive portions of the article being treated must be stopped with wet asbestos in order to prevent the quenching liquid from penetrating into them. In some cases, to slow down the cooling rate in the article edges, where quenching cracks usually begin to develop, special fixtures are used, e. g., for thin-walled, conical parts, as shown in Fig. 97.

In hardening flat, thin articles such as saw discs and disc milling cutters, their distortion can hardly be avoided even when all the regulations pertaining to the cooling and heating operations are carefully complied with. It is recommended, therefore, to quench such parts and tools in specially designed devices in the following way: the heated part is inserted into such a fixture and, upon being quick-

ly clamped in it, is plunged together with it into the quenching tank. It is true that in such method of quenching the resultant hardness of the metal is somewhat lower; but, on the other hand, distortion is practically eliminated. In the mass production of parts or tools liable to distortions in quenching special quenching presses are used, in which the articles to be hardened are quenched after being firmly clamped, which eliminates distortion (warpage). Such quenching presses are used for hardening cone gears, automobile crankshafts, camshafts, cardan shafts,

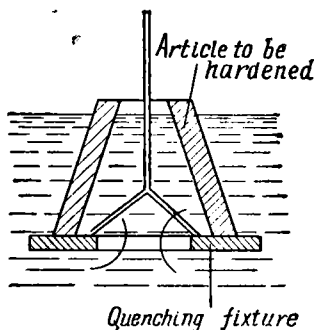


Fig. 97. Sketch illustrating the quenching fixture and method of quenching to be used for preventing cracks developing from the corners or edges of the thin-walled conical article being hardened (as given by A. N. Rozanov).

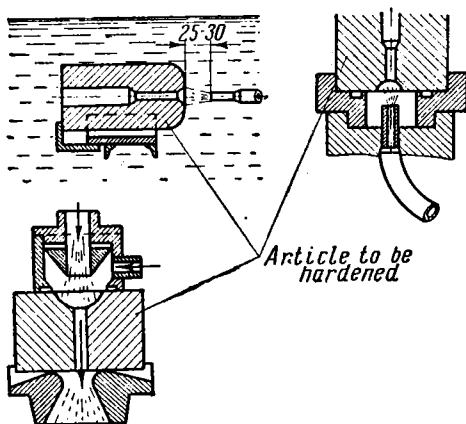


Fig. 98. Sketches illustrating the method of quenching dies by spraying them with water (as given by Y. R. Rauzin).

rear semi-axes, agricultural scythes, ribbon saws and laminated or leaf springs.

Very often in the heat-treating practice the so-called *local hardening* is used, usually for parts that need not be hardened over the entire surface. For example, such article as chisels, bits, cutting tools, centres for lathes and many other parts and tools are always subjected to the local hardening operation, consisting in the quenching of their working parts. The local hardening may be performed in two manners—either by heating the part of tool throughout with the subsequent plunging into water of only the portion to be hardened, or by heating (e. g., in a salt bath) only the portion to be hardened, with subsequent quenching of the whole article.

The so-called quenching by means of *jets* of cold water is to be considered as a particular aspect of the local hardening technique. This method is performed as follows. The heated article or tool is placed in a specially designed fixture, whereupon one or several jets of cold

water are directed against the surface of the portion to be hardened. Such a procedure is completely indispensable in heat-treating holes, e. g., in dies with holes (Fig. 98), drawing die holes, internal surfaces of bushings, etc.

42. VOLUMETRIC CHANGES THROUGH HARDENING

In hardening to form a martensite structure there is always a volumetric expansion of the part or tool being treated, which is explained by the fact that the specific volume of the martensite is greater than that of the ferrite-carbide aggregate. The volumetric expansion of the metal increases with the increase of carbon content. The diagram of Fig. 99 (compiled according to the data of M. G. Oknov) shows the change in the specific volume of carbon steels being hardened to form a martensitic structure.

These volumetric changes are in general not excessive. For the V10 grade steel the volumetric change as a result of hardening to form martensite does not exceed 1 per cent. This means that the linear dimensions of the article increase by about 0.3 per cent, as is evident from the following calculation (assuming the part to have the form of a cube with the side a):

$$a_{\text{quench}}^3 = 1.01a_{\text{initial}}^3, \quad (17)$$

whence it follows that:

$$a_{\text{quench}} = a_{\text{initial}} \sqrt[3]{1.01} = 1.003a_{\text{initial}} \quad (18)$$

where a_{quench} —the length of the cube side after quenching, a_{initial} —the length of the cube side before quenching.

The above data are illustrative of the changes in volume and linear dimensions of articles subjected to the through hardening operation. When the hardening is not of a through type, changes in volume and linear dimensions of the articles treated are correspondingly smaller. In surface hardening the volumetric and linear changes are practically nil.

The hardening temperature produces twofold and diametrically opposite effects upon volumetric changes. On the one hand, the high-

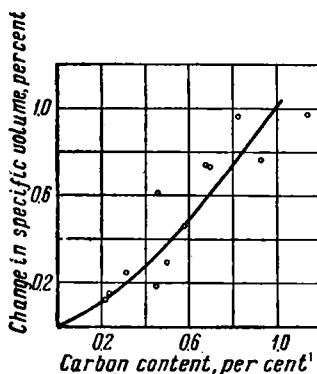


Fig. 99. Diagram showing the change in specific volume of carbon steels being hardened for martensite to form.

er the hardening temperature, the larger is the amount of residual austenite retained in the steel structure and the less the magnitude of volumetric changes, since the specific volume of austenite is less than those of the martensite and ferrite-cementite conglomerate.

Thus, it follows that for carbon steels with a small amount of residual austenite after quenching some decrease in volumetric changes can be brought about through hardening to as low temperatures as possible. For alloy steels, however, in the structure of which a considerable amount of residual austenite is retained the hardening temperature should be somewhat increased in order to decrease the magnitude of volumetric changes.

The changes in volume and linear dimensions that take place in hardening are generally of a very insignificant size and do not produce any substantial effect upon most articles and tools treated. But there are tools for which even insignificant changes of dimension on hardening are entirely inadmissible. These tools are long screw taps, composite dies and punches of blanking stamps and bending stamps, various stamps and other intricate tools, the dimensions of which after hardening cannot be corrected by grinding.

Such tools should be made exclusively from alloy steels; e. g., edged tools must be fabricated of steels of grades XΓ and XBF, and dies and punches of the X12 grade steel and similar steels. Tools made of these steels, being subjected to a special heat-treatment (the so-called *hardening without deformation*), hardly change their original dimensions.

Now let us consider the essentials of the process of hardening without deformation.

It is known that the martensite present in the structure of hardened steel increases the volume of a tool, whereas the presence of residual austenite decreases this volume. It is obvious that at some definite ratio of the martensite and residual austenite present in the steel structure neither a decrease nor an increase in the volume and linear dimensions of a tool will occur, i. e., the dimensions of the tool after quenching and tempering will be exactly the same as they were before hardening.

A. P. Gulyaev recommends the following procedure for hardening tools of the XΓ grade steel without deformation: quenching from a temperature of 890°C into a saltpeter bath heated up to a temperature of 200°C; holding there for 30 minutes and then cooling very slowly (in a cooling down tempering oil bath, or in sand, ashes, etc.). The purpose of increasing the hardening temperature is plain from what has been previously stated. Broken hardening removes the internal stresses and strains that may cause change in the dimensions of articles as a result of their warping. But the slow cooling from the

temperature of 200°C combines the last stage of the hardening operation (the formation of martensite) with tempering. At these temperatures the first transformation in tempering takes place, i. e., the tetragonal martensite is transformed into tempering martensite. Instead of slow cooling from a temperature of 200°C, the metal may be cooled more rapidly (e. g., in still air) with subsequent tempering at between 180 to 200°C.

Another method of quenching without deformation is that of subjecting the steel to a preliminary heat-treating before the final quenching. This consists of quenching with subsequent drawing (tempering at a high temperature). As a result the finely dispersed structure called sorbite is formed, and therefore in heating for the final hardening the austenite becomes more uniform and more highly alloyed. This increases its stability, and consequently a greater amount of residual austenite is retained in the steel structure after final quenching. As a result of such a treatment changes of dimension will be less.

To ensure constant dimensions of tools made of the XBF grade of steel, the following heat-treating operation may be used:

- 1) preliminary quenching in oil from a temperature of 860°C;
- 2) drawing at a temperature of 650°C for 1.5 to 2 hours;
- 3) final quenching in oil from a temperature of between 830 and 840°C;
- 4) tempering at a temperature of 180°C for a period of 1.5 to 2 hours.

For tools of high-chromium steels of grades X12M (standard) and X12Φ (non-standard) dimensions may be kept constant by means of the so-called *finish heat-treating operation*, proposed by Y. R. Razin and E. Liberman. Essentials of this method are: in the structure of a hardened steel there is a quite high proportion of residual austenite, which leads to a reduction of the dimensions of the tool immediately after quenching. The subsequent short-time tempering operations are intended to produce gradual transformation of the residual austenite into martensite, with a simultaneous increase in the dimensions of the article.

The finish heat-treatment consists of the following operations:

- 1) heating the metal to 1050° to 1060°C;
- 2) transferring the article to another furnace heated up to 450°C;
- 3) equalising the temperature of the article and holding it at that temperature for a period of 40 to 50 minutes;
- 4) cooling it in air;
- 5) repeated high tempering at a temperature of 520°C for a period of 40 to 50 minutes each time. After each tempering the dimensions of the article are measured. The tempering is stopped as soon as dimensions measured are found to be equal to the initial ones.

43. HARDENING DEFECTS

When steel articles are hardened many defects may be caused in a number of ways. The principal ones are: 1) oxidation and decarburisation; 2) quenching cracks; 3) distortion and warpage; 4) change in dimensions; 5) mechanical properties obtained not conforming to specifications; 6) soft spots.

Oxidation and decarburisation were dealt with in detail in Chapter V of this book. Causes of the development of cracks, distortion and change in dimensions of articles have already been discussed in paragraphs 40 to 42 of this chapter. To what has been said about the causes of distortion, the following should be added. Distortion is often found to be caused not by the cooling conditions but by the conditions of heating. If the bottom of a heating furnace is uneven, then long, slender articles (e.g., blades for guillotine shears) placed on this bottom bend under their own weight on heating and, of course, remain distorted after hardening. Often they cannot be straightened after hardening. Therefore such parts and tools should be heated on a perfectly even bottom or preferably vertically or suspended in a pit-type heating furnace. Springs, however, should not be heated vertically, or they will inevitably be compressed or extended, with their pitch changing its value. Therefore springs must be heated horizontally, placed on the even bottom of a heating furnace.

Most hardened articles and tools are checked by testing their hardness, and some machine parts are tested to determine whether their mechanical properties conform to specifications. Checking articles by tests of hardness and mechanical properties is usually done only after tempering, but often the causes of unsatisfactory hardness or mechanical properties are found to originate from the hardening procedure. The most frequent defect in hardened tools is too low a hardness; this kind of spoilage is known to be caused by the following factors:

1) Insufficiently rapid cooling, e. g., because the quenching liquid is either overheated or polluted; also because of too hasty transfer from water into the oil bath (when quenching through water to oil).

2) Short holding at the hardening temperature, which is insufficient for the globular pearlite to transform completely into austenite. In this case the hardened steel consists of martensite and globular pearlite (Fig. 100).

3) Low hardening temperature, which may be the consequence not only of a too low or non-uniform temperature in the heating furnace where the tool treated has been heated, but of a considerable drop in the temperature of the tool during its transference from the furnace into the quenching tank, possibly caused by an unduly long distance

between the two (such cases can still be met with in practice in small heat-treating workshops where the equipment layout has not been rationally planned).

4) Decarburisation of the tool surface.

Sometimes cases of spoilage because of non-uniform hardness on the tool surface are reported. Such tools are found to have hard and soft areas, the latter being called "soft spots". This may be caused either by the presence of portions with decarburised surface or by the tools not having been thoroughly moved about in the quenching liquid, with an uneven cooling resulting. With insufficiently vigorous or irregular (Fig. 101) agitating of the tool in the quenching bath, the tendency to formation of a vapour coating on certain surfaces may increase, leading to an abrupt drop in the cooling rate. Small

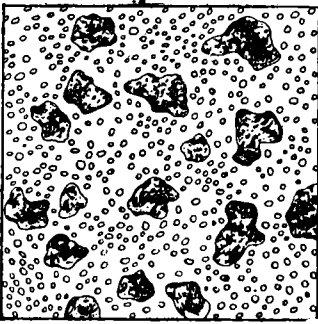


Fig. 100. Structure of V8 grade steel which has been held for an insufficient period of time at the hardening temperature. Martensite and globular pearlite, 500 $\times$.

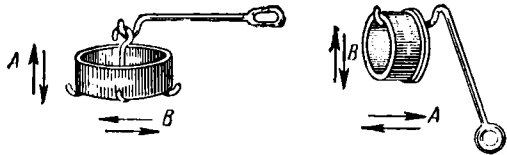


Fig. 101. Sketches illustrating the correct A and wrong B methods of the moving parts to be hardened in the quenching liquid (as given by Y. R. Rauzin).

soft spots can be prevented by adding a small amount (2 to 3 per cent) of soda ash to the water. When the water evaporates on the surface of the tool small soda crystals are depositing; these instantly fly off it, breaking up the vapour coating. When quenching is performed by spraying with water soft spots are completely eliminated.

Causes preventing the mechanical properties obtained from conforming to specifications will be discussed in the chapter on tempering (see Chapter VIII).

To conclude this short survey of defects in hardening special attention should be called to the fact that in analysing causes of a spoilage the initial material must be carefully checked. Cases of spoilage may be reported where it is explained simply by the fact that a part or tool has been fabricated from steel of the wrong grade, and defects may also be caused by faults in the initial material.

44. PECULIAR FEATURES OF THE HARDENING OF HIGHLY ALLOYED STEELS OF THE AUSTENITE AND CARBIDE CLASSES

Many steels of the austenite class in as-annealed condition consist of grains of austenite and carbides (Fig. 102). Carbides affect the physical properties of austenitic steels. Firstly, they may decrease its plasticity and impact strength and their presence in the steel structure is therefore undesirable. But this is not the main point. They are chiefly undesirable because part of the alloying elements remain in them, thus depleting the austenite of these elements and making it less resistant to scaling, acids and wear. From the point of view of acid and corrosion resistance in general, carbides are undesirable in the steel structure also for the following reason. As is known from elementary theory on the corrosion of metals, the best corrosion resistance is found (under otherwise equal conditions) in alloys with a uniform structure. The presence in the structure of two or several phases gives rise to the formation of numerous microelements, which greatly accelerate corrosion. Therefore

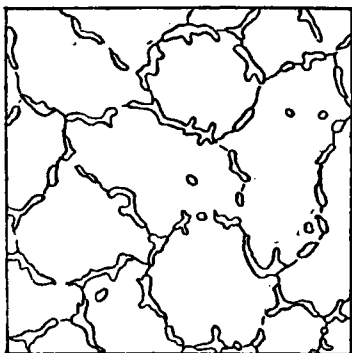


Fig. 102. High-manganese steel of grade T13 (Hadfield steel) in as-cast condition, 300 $\times$.

a steel with purely austenitic structure possesses greater resistance to corrosion than the same steel with a structure consisting of austenitic grains and small grains of carbides.

Thus, the presence of carbides in the structure of a steel of the austenitic class is nearly always undesirable and very often highly detrimental. But how can these carbides be removed from the steel structure? The only method is to dissolve them in the austenite. For the carbides of most elements this method is quite practical, as the solubility of carbon and alloying elements in austenite increases as the temperature rises. Therefore the heat-treating process applied to steels of the austenitic class may be generally represented as the following sequence:

- 1) heating the steel to a temperature at which all, or at least most, of the carbides dissolve in the austenite;
- 2) quenching in water in order to prevent the separation of carbides from the austenite, which takes place on slow cooling. The structure of hardened steels of the austenitic class (e. g., of high-manganese or Hadfield steel) is found to consist entirely

of the austenitic grains (Fig. 103) even when their carbon content is high.

Parts of heat-resisting austenitic steels usually work at temperatures from 500 to 650°C. This gives rise to some peculiar features in the heat-treating of these steels. The heat-treating process performed on heat-resisting austenitic steels consists in hardening at high temperatures (1000 to 1100°C) followed by stabilising or ageing. As a result of hardening a uniform austenitic structure develops. Stabilisation for a period of several hours is conducted at temperatures

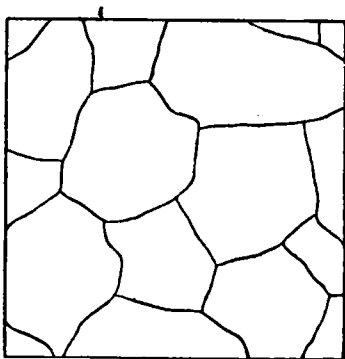


Fig. 103. Microstructure of high-manganese steel of grade F13 (Hadfield steel) in as-hardened condition, 300 $\times$.

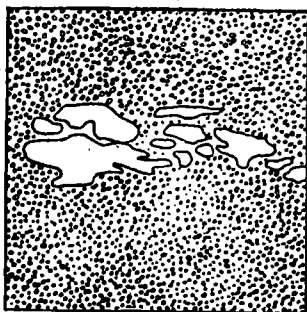


Fig. 104. Carbide segregation as revealed in the structure of high-speed steel, 500 $\times$.

of 600 to 850°C, thus exceeding the working temperatures when in service. In the stabilising operation the finely dispersed carbide particles separate from the austenite, while in steels alloyed with titanium and other similar elements particles of intermetallic compounds, i.e., compounds consisting of two or more metals (e.g., Ni, Ti), also separate. During stabilisation a process of coagulation or coalescence of carbides and intermetallic compounds takes place until they reach a certain size, after which the growth of particles ceases. In any case, these particles will not grow at the working temperatures of steel. This determines the stability of the structure and heat-resisting properties of heat-resistant steels of the austenitic class when in service at elevated temperatures.

Most steels of the *carbide class* are classified as tool steels. In as-cast conditions (see Fig. 7), the ledeburite carbides are distributed between grains of the matrix in the form of a ledeburite eutectic. Forging or some other hot-working operation will break down the ledeburite eutectic with carbides taking a globular form (see Fig. 8).

As a result of hot forging on plain anvils ledeburite carbides sometimes become distributed in strips (Fig. 104). This phenomenon is called carbide non-uniformity or carbide segregation, and is a highly detrimental factor. Tools of steels of the carbide class with a marked carbide non-uniformity are rapidly blunted in service. The carbide segregation cannot be removed through any heat-treating operation, since the ledeburite carbides cannot be dissolved in the austenite.

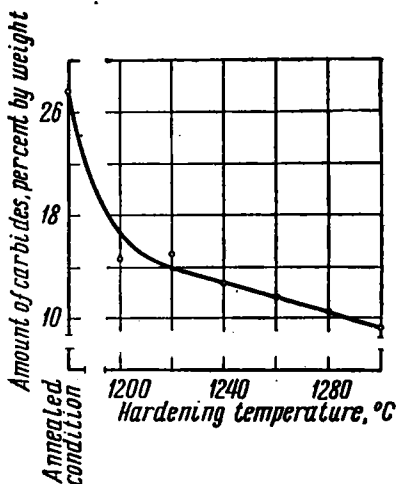


Fig. 105. Diagram showing the effect of the heating temperature for hardening high-speed steel upon the amount of carbides undissolved in the austenite of the steel structure (as given by A. I. Gordin).

The structure of steels of the carbide class in as-forged and as-annealed states consists of sorbite or globular pearlite and secondary and ledeburite carbides (see Fig. 8). Steels of the ledeburite class, like steels of the pearlitic class, are capable of undergoing phase transformations, and consequently they may be hardened for the formation of a martensitic structure. The large amount of carbides present in the structure of these steels makes it necessary to modify somewhat the ordinary heat-treating process. The peculiar features of the heat-treating operation applied to such steels will be considered below.

The most typical steels of the carbide class are the high-speed steels. Therefore the peculiar features

of heat-treating steels of the carbide class will be considered for high-speed steels only, especially as they are of a great practical importance.

The remarkable property of high-speed steels is their red or secondary hardness, i. e., their ability to retain great hardness and cutting power at high temperatures, up to about 600°C. This is indicative of the high thermal stability of martensite. The question arises as to why the martensite in high-speed steels is more heat-resistant than, say, the martensite of carbon steels. What is the difference between them? There is a very simple answer: the martensite of high-speed steels is alloyed (and to a considerable extent) with such elements as tungsten, chromium, and vanadium, which make its space lattice more stable, preventing it from precipitating carbon atoms, i. e., they prevent the martensite from decomposition.

An increase in the content of alloying elements for martensite is achieved in the same way as for steels of the austenite class, i. e.,

through heating to high temperatures for hardening. High-speed steels are heated to temperatures near the fusion points of the former, as:

Grade of steel	Hardening temperature, °C
P18	1250 to 1280
P9	1220 to 1240

The higher the hardening temperature, the greater the amount of carbides dissolved in the austenite (Fig. 105) and the more highly alloyed it results. But it is impossible to dissolve all the carbides in the austenite in heating, because their amount exceeds the solubility limit (it should be remembered that the steel belongs to the ledeburite class). In subsequent hardening the austenite partly transforms into martensite, and the structure of the resulting hardened high-speed steel consists of martensite, residual austenite, and undissolved carbides. The subsequent tempering is apt to transform the greater part of the residual austenite into martensite.

In addition to the high-speed steels, steels of the carbide class include also high-chromium steels for dies, for which the following optimum hardening temperatures have been specified:

Grade of steel	Hardening temperature, °C
X12	975 to 1000
X12M	1020 to 1030
X12Φ	1020 to 1030

Chapter VIII

TEMPERING AND SUB-ZERO TREATMENT

45. THE OBJECT OF TEMPERING

Hardened machine parts and tools are usually subjected to tempering. Exception is made only of parts made of steels of the austenitic class, which as a rule are not tempered after hardening. In some cases the tempering operation is performed even after normalising.

Hardened steel should be tempered for the following reasons.

The structure of hardened steel consists mainly of tetragonal martensite and a certain amount of residual austenite. In a hardened high-carbon steel there are also carbides that are not dissolved in the austenite when heated for quenching. The properties of hardened steel are largely determined by the properties of the tetragonal martensite, *the hardness of which is very high*. The structure of the hardened steel, consisting in the main of tetragonal martensite, would

therefore be quite suitable for tools if the great hardness of the tetragonal martensite were not combined with very *low ductile properties* and a very *poor impact strength*.

Because of this hardened tools must be tempered. It is known that in tempering the tetragonal martensite gradually changes to a heterogeneous mixture consisting of the alpha solution being depleted of carbon and with a decreasing degree of tetragonality, and highly dispersed carbides.

As a result of this primary transformation, on the one hand, some decrease in hardness takes place, and, on the other hand, there is a progressive increase in the plasticity and impact strength.

Various degrees of increasing the plasticity and impact strength of the steel by tempering have to be adopted, according to designation of different tools. For cutting tools, the cutting properties of which are substantially deteriorated by even a slight decrease in the hardness in tempering, the minimum tempering temperature capable of removing the brittleness is selected. In practice it has been found sufficient to temper cutting tools of carbon steels at temperatures as high as 160° to 200°C. Tools subjected to shock (e. g., erecting and fitter's tools, dies, punches, and broaches) for which a somewhat greater decrease in hardness may be allowed but for which impact values obtained on tempering at 160° to 200°C are quite insufficient, are to be tempered at higher temperatures, e. g., 200° to 300°C.

The structure of most machine parts of structural steels should be sorbite, as it possesses the *best combination of mechanical properties—high tensile strength, elasticity, plasticity and impact strength*.

Tempering has been defined as *reheating to temperatures up to 400°C*, corresponding to the temperatures of the first, second, and third transformations in steel, and high tempering (or drawing) as *reheating to higher temperatures*, at which occurs the fourth transformation (recrystallisation of ferrite, coagulation or coalescence of cementite, and the formation of troostite and sorbite).

46. TEMPERING AT LOW TEMPERATURES

As has been already pointed out, tempering at low temperatures, briefly called "low tempering", is applied to different tools. Table 17 shows tempering temperatures for main groups of tools made of steels of the pearlitic class.

Low tempering of tools is carried out either in oil baths (up to a temperature of 250°C) or in saltpeter baths. The tempering may be also conducted in furnaces with air atmosphere provided that this is made to circulate by some artificial means. At comparatively low

Table 17

Tempering Temperatures for Main Types of Carbon Steel Tools

Kind of tool	Grade of steel	Tempering temperature, °C	Hardness, Rockwell C
Cutting tools, such as milling cutters, bits, screw taps	V10, V11, V12	150 to 220	60 to 64
Dies and punches for swaging or cold forging	V10	220 to 260	55 to 59
Dies and punches for cold stamping of sheet metal . .	V11, V12	250 to 270	56 to 60
Tools subjected to shock (chisels, etc.)	V7	280 to 300	53 to 56
Hammer dies	V7	350 to 450	302 to 444 H_B (Brinell hardness number)

temperatures (up to 500° to 600°C) the heat is transmitted from the still air to the metal treated very slowly and, which makes the matter worse, very unevenly. Very frequently a tool held in the tempering furnace for a specified period of time proves to be not fully tempered if the atmosphere in the furnace heating chamber is not made to circulate, and the tool thus tempered usually has a very short service life. Insufficient attention is often paid to this fact in heat-treating practice. Unfortunately the microstructure cannot be used as a basis for determining the degree of low tempering, for the normally tempered tool and the tool being not fully tempered both have absolutely the same microstructures. It is true that the higher the tempering temperature, the greater is the degree of etching of martensite, but this factor is very indeterminate one, as the etching capacity to a very large extent depends on the duration of etching.

It is extremely important that the steel be tempered *as soon as possible after quenching*. In heat-treating practice there are cases reported, where tools, especially those of intricate shape, which have been satisfactorily quenched without any cracks forming on cooling, suddenly develop cracks while lying undisturbed on stands. This is the result of the severe internal stresses set up by quenching. Therefore such tools should preferably be put into a tempering furnace immediately after quenching.

Sometimes small tools of regular design are tempered by the heat left in the interior portion of the article, when the cooling has not been allowed to go to completion. Such heat-treating has been termed *self-tempering* and is performed as follows. The tool is with-

drawn from the quenching tank where it has been held before cooling is complete, and a small area is rapidly ground on its surface by means of a file or emery paper. Since the tool has been not allowed to cool completely, the heat maintained in its central portion will gradually reheat its quenched surface, with the result that the temper colours will appear on the polished surface. When the colour of this surface eventually changes to the desired temper colour, the tool is once more plunged into the quenching tank for final cooling in order to prevent the further rise of its temperature.

Self-tempering is usually applied to tools that are subjected to local hardening, such as chisels. Chisels are immersed in the quenching bath with their working parts only, the heat in the unquenched parts being made to temper the hardened portions. Hardening with subsequent self-tempering is used only in small tool-making and maintenance shops in treating small lots of tools.

Temper colours develop on the metal surface as the heat increases, due to the formation of an extremely thin, transparent film of oxide. In this surface film occurs the interference of light, which gives rise to the appearance, according to the film thickness, of some temper colour. The method of determining tempering temperatures by colours is based on the fact that each gives rise to a certain thickness of oxide film which, in turn, determines the change of colour.

Tempering colours	Temperatures, corresponding to them, °C
Pale yellow	220
Straw-yellow	240
Yellow-brown	255
Spotted red-brown	265
Light purple	275
Dark purple	285
Dark blue	295
Pale blue	315
Gray	330

At temperatures above 330°C the tempering colours vanish and the surface film of oxide becomes opaque. In determining temperatures by the colour method it is necessary to take into account that the temper colours correspond to the above-mentioned temperatures only at the instant of their formation or when held for only a short period of time (up to 2 to 3 minutes); if the given temperature is maintained for a longer period of time, however, the temper colour will undergo a change into another colour characteristic of a higher temperature. The indicated relationship between the temper colours and temperatures is valid only for plain carbon and low-alloy steels. For highly alloyed steels and non-ferrous alloys there should be different relationships. However, up to now no verified and reliable data for them are available.

In some cases an *additional local tempering* is conducted besides the main one. This heat-treating operation becomes necessary in cases where one part of a tool should be less hard than another. For example, in hammer dies shanks should be softer than half-impressions surfaces, and the hardnesses of shanks of bits, screw taps and reamers must be also lower than those of their working parts. There are cases where this can be achieved by local hardening the working parts of tools alone; naturally shanks will not be hardened in such an operation, so that the hardness of these parts will be lower both after hardening and annealing. But often local hardening is either difficult or even impracticable (e. g., as in the case of dies), and in such cases additional local tempering has to be used. For this purpose large tools such as dies are put into the heating furnace only partially, i. e., not as a whole (Fig. 106). For additional tempering of small tools they are partially immersed in a salt bath. The shank of a tool can acquire the required hardness through an individual selection of the bath temperature, as well as of the duration and degree of immersion.

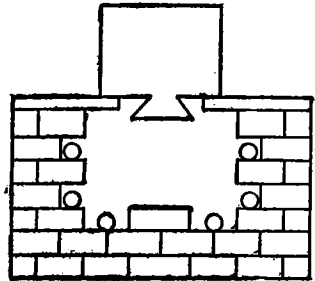


Fig. 106. Sketch illustrating local tempering of a die shank.

47. TEMPERING AT ELEVATED TEMPERATURES (HIGH TEMPERING)

Machine parts are usually drawn-back after hardening. This operation, called "high tempering", is performed to give them a certain *combination of strength with ductile and impact properties*. The criterion for the tempering of steel at elevated temperatures may be either the specified hardness (the value most easily determined), or the given mechanical properties.

The *temperature* is the most important factor of the high tempering operation, as it is for any heat-treating process. For most tools subjected to low tempering, the temperature of the process is fairly definitely ascertained, so that there is no necessity to select it each time because the kind of tool to be treated determines the hardness desired and therefore the tempering temperature, too. For machine parts there is no such certainty because of the great varieties of their nomenclature and hardness values required. *Their high tempering temperatures* have to be selected for each individual case and must be indicated in process charts.

To determine the temperature and duration of the tempering process, a nomograph, compiled by Professor A. P. Gulyaev, may be used (Fig. 107). This nomograph consists of two diagrams, the left one showing the relation of the 0.35 per cent carbon steel hardness to the holding time of heating for tempering, and the right one the relation between the hardness and the carbon content of the steel.

This nomograph should be used as follows. Let us suppose that we have to determine the temperature and holding time of heating for tempering the Y8 grade steel so as to obtain the hardness 45 Rockwell C. On the right diagram is found point 1, corresponding to the

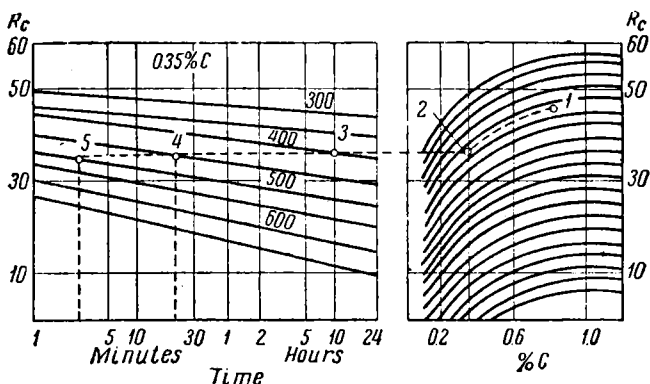


Fig. 107. Nomograph to be used for determining the tempering temperature and duration of the tempering process.

above data. From point 1 a curve, parallel with the curves on the right diagram, is to be drawn to point 2 in the line corresponding to the carbon content of 0.35 per cent, i. e., to the sort of carbon content for which the left diagram has been drawn up. From point 2 a straight line is to be drawn toward the left side, parallel to the abscissa axis. This straight line will intersect a number of temperature lines on the left diagram at such points as 3 (corresponding to the tempering temperature of 400°C), 4 (corresponding to the tempering temperature of 450°C) and 5 (corresponding to the tempering temperature of 500°C). By projecting points 3, 4, and 5 on the axis of abscissa, the holding times of heating for tempering corresponding to these temperatures may be found. This means that to obtain hardness of 45 Rockwell C in Y8 grade steel it should be tempered:

- at 400° C for a period of 10 hours;
- or at 450° C for a period of 20 minutes;
- or at 500° C for a period of 3 minutes.

Tempering for a very short period of time is impractical because of difficulties in controlling precisely the time required for the operation. The nomograph gives only the holding time at a given temperature and does not take into account the time of heating up to this temperature, but at comparatively low temperatures, at which the articles treated do not glow with red heat, it is practically impossible to ascertain when the heating is completed and the holding begins. Tempering for a long time is also undesirable from the production standpoint. For the given example, therefore, the best conditions of tempering should be holding at 450°C for 20 minutes.

The nomograph may be applied to plain carbon steels and those alloy steels that are alloyed with

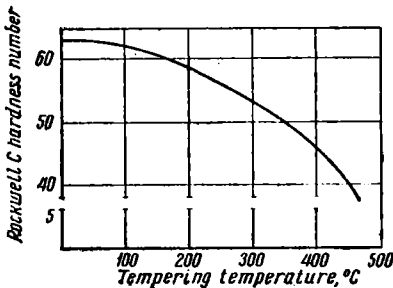


Fig. 108. Diagram showing the relation of 65Γ grade steel hardness to the tempering temperature.

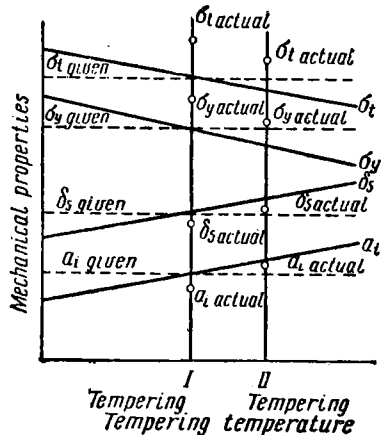


Fig. 109. Diagram illustrating the adjusting of the repeated tempering temperature to mechanical testing results.

only one element and contain not more than 1 per cent of chromium, silicon or tungsten, or not more than 0.5 per cent molybdenum.

To determine the temperatures of the high tempering on the basis of the hardness specified diagrams of the type represented in Fig. 108 for the 65Γ grade steel may be used.

The procedure of determining the high tempering temperatures on the basis of the specified mechanical properties is of a more complex character. To ascertain these temperatures the technologist must have at his disposal a diagram of the type shown in Fig. 46. By using such a diagram the technologist will be able to select the desired tempering temperature, at which the mechanical properties, as determined from the diagram, will be above the specified values.

After hardening and tempering have been completed special proportions known as coupons, which are attached to the article treated, are cut off, prepared for testing and finally tested for mechanical

properties. The results obtained in testing such specimens usually correspond to specifications.

Sometimes, however, the actual values for some set of properties prove to be lower than the specified values, e. g., either 1) actual values for tensile strength and yield point may exceed the specified values, while test results for elongation, reduction of area and impact strength are well below them (Fig. 109, *I*); or 2) on the contrary, the test results for tensile strength and yield point may be lower than the specified values with test results for elongation, reduction of area and impact strength being above the given values.

The first case gives evidence of a tempering temperature having been selected too low, and the second of one too high. To make test results correspond to the specified values in the first case, it is necessary to conduct an additional tempering (Fig. 109, *II*) by reheating the steel to a somewhat higher temperature, while in the second case the article has to undergo a full repeated heat-treatment consisting of hardening followed by tempering at a somewhat lower temperature.

In hardened articles considerable internal stresses are set up by the quenching operation. Therefore articles to be tempered are charged either into cold furnaces or, at least, into furnaces cooled down to 200 to 300°C, held there at that temperature for several hours for the temperature of the steel to become equalised, and then slowly heated at a rate of 50 to 100°C per hour (according to the size of cross-section) to the given tempering temperature. Tempering by slow heating in a furnace is the most important case in the heat-treating practice where slow heating is actually indispensable (see Chapter V).

There are two ways of cooling from the tempering temperature, depending upon the kind and grade of steel: 1) slow cooling in a furnace; when the article is made of steels not subject to temper brittleness, or 2) rapid cooling in water or oil, when the article is made of steels subject to temper brittleness (for the list of both types of steels see Chapter IV).

48. PECULIAR FEATURES OF THE TEMPERING OF ALLOY STEELS OF THE CARBIDE CLASS

There are some peculiar features in tempering tools of steels of the carbide class as compared with those of pearlitic steels.

These peculiarities are explained as follows. The higher the quenching temperature, the more highly alloyed becomes the austenite and the greater the amount of residual austenite retained in the structure of the hardened steel. This residual austenite, being of course also highly alloyed, possesses high stability at elevated temperatures. The diagram in Fig. 110 shows the relation of the hardness of X12

grade chromium steel to the hardening and tempering temperatures. To obtain high hardness in a tool quenched from a high temperature, tempering at higher temperatures has to be employed also, as is evident from Fig. 110. The question may arise as to why the steel should be quenched from such high temperatures if the high hardness would be also obtained when quenching from lower temperatures (e. g., see curves in Fig. 110). This is explained by the fact *that not only hardness matters*. High-speed steels quenched from a low temperature will acquire a very high hardness but they will be lacking in the most valuable of their properties — *red hardness*. Though dies of type X12 high-chromium steels quenched from a low temperature will have high hardness, they will prove to be less strong and tough than the same dies hardened from higher temperatures. And at low hardening temperatures a small amount of austenite will be retained in the structure of these steels, so that they will not respond to the finishing heat-treatment in order to correct their dimensions (see Chapter VII). Yet the response to the finishing heat-treating operation is one of the most valuable properties of these steels.

Therefore tools of highly alloyed steels of the carbide class, quenched from elevated temperatures, should be tempered at temperatures such as:

Grade of steel	Temperature, °C	
	Hardening	Tempering
P18	1250 to 1280	550 to 560
P9	1220 to 1240	
X12	975 to 1000	500 to 550
X12M	1020 to 1030	
X12Φ	1020 to 1030	

The question may also arise as to whether martensite transforms into troostite at such high tempering temperatures. It does not do so, because the martensite is also highly alloyed and its heat-stability markedly increased.

The second peculiar feature of the tempering of highly alloyed steels of the carbide class is that the austenite transforms into martens-

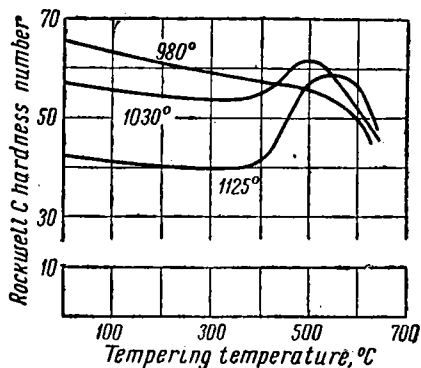


Fig. 110. Diagram showing the relation of the hardness of the X12 grade chromium steel to the hardening and tempering temperatures (as given by Y. R. Rauzin).

ite not at the time of holding at the tempering temperature but *on cooling after holding*. The following interesting data are available, compiled by Y. P. Rauzin: After the X12M grade steel was tempered at 520° C for 2 hours, 28 per cent of the residual austenite in its structure had been decomposed; after it was tempered at the same temperature in two 1 hour stages, 45 per cent of the residual austenite had been decomposed; and after tempering it under the same temperature conditions in four 30 minute stages, 55 per cent of the residual austenite had been transformed into martensite.

The transformation of the austenite during cooling and not at the time of holding may probably be attributed to the fact that on cooling there are set up the internal stresses that alone can suppress the great stability of the austenite. Therefore all tools made of highly alloyed steels of the carbide class are necessarily subjected to the multiple-stage, or repeated tempering operation.

49. AGEING OF HARDENED STEEL

Internal stresses set up by the hardening process are only partly relieved by low tempering (Fig. 111). The greater part remains unaffected, but with time they are released spontaneously. This spontaneous

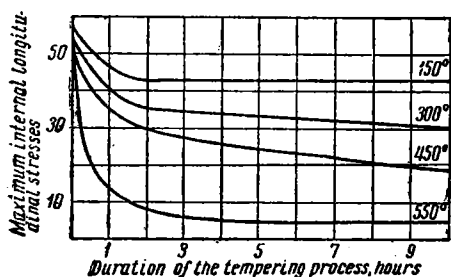


Fig. 111. Diagram showing degrees of relieving hardening stresses at different tempering temperatures (as given by Buchholz, Buhler and Schulz).

release is accompanied by a change in the shape and dimensions of the hardened tool. Such a gradual process of spontaneous changing of shape and dimensions of a hardened tool being tempered at low temperature has been termed the *natural ageing process*.

Apart from the spontaneous release of the internal stresses, another cause bringing about the change in dimensions of the hardened steel is believed to be the isothermal

transformation of the residual austenite, which is not removed from the steel structure on tempering at temperatures below 200° C. Changes in dimensions resulting from the natural ageing of the steel are extremely small, being measured in microns. For the great majority of tools and articles such minor changes in dimensions are of no practical importance and can be simply disregarded. Yet for parts of measuring instruments, and especially for gauges and ball

bearings, even such very small changes in dimensions are absolutely inadmissible.

To prevent tools and articles from changing their dimensions with the lapse of time they are subjected to artificial ageing. This particular type of heat-treatment consists of heating the tools or articles up to 100 to 150°C and holding them at that temperature for 10 to 20 hours or even longer. The artificial ageing of steel is performed either in boiling water or in an oil bath (the latter being chosen when the ageing temperature exceeds 100°C). Heating to the ageing temperature and cooling after ageing must be conducted slowly.

Dimensions are changed by artificial ageing much more rapidly than by the natural ageing process, with the result that a change in dimensions which takes many months or even years to be completed in natural ageing will require only several hours. After artificial ageing the dimensions of tools and articles become practically stable. Of course, some minor changes in dimensions do in fact occur even after artificial ageing, but they are so small that they can be completely disregarded.

Instead of the long process of artificial ageing, which sometimes takes many hours, a special method of hardening capable of ensuring stable dimensions has been proposed. The peculiar feature of this process is that after ordinary quenching in water at room temperature (for carbon steels) or in warmed oil (for chromium steels) the metal is additionally cooled in cold water at a temperature near to zero, or preferably the hardened articles and tools are cooled to 10°C. Then an ordinary tempering operation is conducted.

The essence of the hardening process for ensuring stable dimensions is that by cooling articles down to a temperature of between zero to -10°C the residual austenite is practically removed from the structure of the hardened steel, as the carbon content in the martensite of hypereutectoid steels hardened at ordinary temperatures does not usually exceed 0.5 to 0.6 per cent when the M_s temperature is near to zero. Since the residual austenite is thus eliminated, this removes one of the causes of the change in dimensions of articles treated.

The heat-treating operation described above may thus considerably reduce the dimensional changes, making the artificial ageing of the steel unnecessary.

50. SUB-ZERO TREATMENT OF STEEL

Many cases are reported where a certain amount of the residual austenite, often a very considerable amount, still remains in the structure of a hardened steel. In most cases this residual austenite is considered as an undesirable phenomenon, for the following reasons:

1) As the austenite is much less hard than martensite, the hardness of a hardened steel containing residual austenite is well below the maximum; 2) as has been just pointed out, the residual austenite brings about a gradual, spontaneous change in dimensions of articles treated; 3) as the austenite is non-magnetic, its presence in steel magnets decreases the magnetic induction and, consequently, the lifting power of magnets.

The residual austenite remains in the steel structure because for many steels the austenite completes its transformation into martensite at temperatures well below zero (see Fig. 26). This suggests a simple method for removing residual austenite from the structure of a hardened steel, consisting of cooling it to a temperature well below zero, i. e., of subjecting it to the sub-zero treatment. This is the method proposed by Professor A. P. Gulyaev in 1937.

The temperature of the sub-zero treatment depends on the posi-

Table 18

Temperatures of Martensitic Transformation and the Effect of Sub-Zero Cooling on Some Standard Steels Quenched from a Temperature of between 1000 to 1100° C

Grade of steel	Temperature of the martensitic transformation, °C		Amount of residual austenite, in per cent		Increase in Rockwell C hardness after cooling to M_e point
	Beginning (M)	End (M_e)	After cooling to +20° C	After cooling to M_e	
Y7	300 to 255	—55	up to 5	up to 1	up to 0.5
Y8	255 to 230	—55	3 to 8	1 to 6	up to 1
Y9	230 to 210	—55	5 to 12	3 to 10	1 to 1.5
Y10	210 to 175	—60	6 to 18	4 to 12	1.5 to 3
Y12	175 to 160	—70	10 to 23	5 to 14	3 to 4
X05	150 to 140	—95	15 to 30	2 to 14	4 to 7
X09	175 to 150	—85	10 to 27	5 to 14	2 to 4
7X	280 to 230	—55	3 to 10	1 to 8	up to 1
9X	220 to 180	—70	6 to 18	4 to 13	1 to 2.5
7X3	240 to 185	—60	4 to 17	2 to 12	1 to 2.5
X	175 to 145	—90	10 to 28	5 to 14	3 to 6
9XC	210 to 185	—60	6 to 17	4 to 12	1.5 to 2.5
III X9	195 to 150	—85	7 to 27	4 to 14	2 to 5
III X15	180 to 145	—90	9 to 28	4 to 14	3 to 6
XBF	155 to 120	—110	13 to 45	2 to 17	up to 10
XΓ	120 to 100	—120	22 to 60	up to 20	up to 15
60Γ and 70Γ	290 to 230	—55	up to 8	up to 6	up to 1
15X and 20X *	175 to 150	—85	10 to 25	5 to 12	3 to 5
18XHBA *	130 to 120	—110	20 to 45	up to 15	up to 10

* Data refer to the carburised layer.

tion of the M_e point indicative of the end of the martensitic transformation. The steel can be cooled below the M_e temperature, but this evidently achieves nothing, because it cannot bring about any additional increase in hardness since the martensitic transformation ceases at the M_e temperature (even if the amount of residual austenite does not reach zero at that temperature).

Table 18 gives the temperatures of the martensitic transformation, as well as data on amounts of residual austenite and increase in hardness (as given by V. G. Vorobyev).

Most steels can be cooled to sub-zero temperatures in a low-cooling unit with solid carbon dioxide, or dry-ice, with a temperature of -78°C . The low-cooling unit consists of two vessels, the interior one of copper, where parts or tools to be intensely cooled are placed, and the exterior one of steel provided with a good heat-insulation. The space between the vessels is filled with solid carbon dioxide. To attain still lower temperatures other substances are used—the so-called coolants, such as ethane, ethylene, freon-13, etc.

Tools and articles should not be held at sub-zero temperatures for very long; as soon as the part treated has been cooled to the temperature of the low-cooling unit, it may be withdrawn from it. Further holding does not serve any purpose, since, as was previously stated, the austenite is not likely to transform into martensite at constant temperature. It may be accepted that the approximate total time of cooling to a temperature of between -80° to -100°C and holding at that temperature ought to be from $1/2$ to 1 hour.

It is very important to note that the sub-zero treatment should be carried out on the metal *immediately after* its hardening. If there is a considerable lapse of time between the hardening and low cooling treatment the austenite of some steels will be stabilised, and will therefore not decompose on subsequent cooling. Tempering also produces the stabilising effect upon the austenite, as is evident from the data compiled by A. P. Gulyaev and M. S. Chaadaeva.

Grade of steel	Hardening temperature, $^\circ\text{C}$	Increase in percentage of martensite	
		without tempering	after tempering
Y8	780	1.2	0
Y10	780	1.6	0
Y12	780	3.2	0
III X15	850	3.0	0
XI	850	12.7	0
XB	820	4.4	0

There is no advantage in intense cooling of these steels after tempering. Therefore the heat-treating operations should be conducted in the following order: hardening-low cooling-tempering.

The sub-zero treatment has been most extensively used 1) in heat-treating tools of high-speed steel, which in this case should be subjected to only a simple or single-stage tempering; 2) in heat-treating articles and tools, the dimensions of which have to be stabilised (instead of artificial ageing); and 3) in heat-treating carburised articles made from such alloy steels as grades 18XHBA, 12X2H4A, in order to increase their hardness and wear-resistance.

Chapter IX

SURFACE HARDENING

51. OBJECTIVES AND USES OF SURFACE HARDENING

For the great majority of machine parts the high hardness of their surface layers is of special importance. An example is gears, the most heavily stressed part of which is the surface layer of the teeth, since these have to work under conditions of continuous, severe friction. If the surface of the teeth is insufficiently hard they will rapidly wear down and their shape will change, thus making smooth power transmission impossible.

The hardness and wear-resistance of the surface layers of machine parts may be increased in the following three main ways:

1) mechanical surface hardening by shot-blasting or burnishing between rolls; the outer layer of the metal being cold hardened because of this working, its hardness and yield point progressively increase;

2) chemical-thermal treatment (case hardening), such as carburising, cyaniding and nitriding;

3) surface hardening.

Mechanical methods of hardening, which have nothing in common with heat-treatment, will not be considered in this book.

The chemical-thermal treatments will be dealt with in detail in the following chapter, and in this chapter methods of *surface hardening* will be considered.

First we must explain what the advantages of the surface hardening method over an ordinary (through-type) hardening actually are. Ordinary or through-type hardening can also produce a hard, wear-resistant surface layer; or, more strictly speaking, in through hardening the whole article becomes hard and wear-resistant, including its surface layers. In surface hardening, on the other hand, only the surface layer of the article becomes hard and wear-resistant, whereas its core remains comparatively soft and ductile. Why, then, is surface hardening considered to be better than through hardening?

The explanation is that though in through hardening a high hardness is obtained in the whole cross-section of the article treated, at the same time the impact strength of the metal is lower. For the impact value to be increased, the steel must be tempered at high temperatures. In this treatment, the impact strength increases sharply, becoming even higher than that of a non-treated steel, but there is a considerable simultaneous drop in hardness, including that of the surface layer. In articles subjected to surface hardening the interior portion or core, being non-hardened, has lower impact strength, while the surface hardness is much higher (Fig. 112).

Certain articles must have a high impact value combined with a high strength and elastic properties across their whole cross-section. Such articles are, for example, heavy-duty axles, shafts and gears, transmitting heavy torques and working under conditions of dynamic loads, as well as springs, etc.

Such machine parts cannot be given the surface hardening treatment, even if this could be realised technically. Yet there are many machine parts and articles which work in conditions

under which the low strength and impact resistance of raw or thermally non-treated steel would be quite sufficient. These may be subjected to surface hardening.

In addition to this, the surface hardening method has a number of advantages from the point of view of technology and production.

Surface hardening does not require heating furnaces and can be performed in any part of a machine-shop. This means that it can be incorporated into the general production line of the machine shop.

Surface hardening process is performed much more rapidly than ordinary hardening: hence the total period of time required for manufacturing articles is considerably reduced.

Surface hardening does not give rise to oxidation and decarburisation of the metal treated.

Finally at least the principal methods of surface hardening, such as induction hardening and electrolytic hardening, can be partially or totally automatised.

Because of the above-mentioned and many other advantages of surface hardening this process is considered a very progressive and advanced method of heat-treatment of steel.

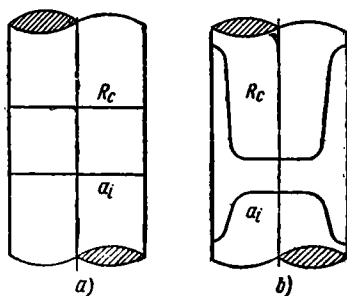


Fig. 112. Sketch illustrating the comparison between the mechanical properties of the same steel article as developed in through hardening (the case a) and in surface hardening (the case b).

The following methods of surface hardening are now used:

- 1) flame hardening;
- 2) hardening in electrolytic bath;
- 3) induction hardening.

Of these the latter, which involves rapidly heating the surface of the steel to the desired depth above the critical point by means of electrical induction and cooling as desired, has unquestionable advantages over the rest. Therefore induction hardening is now widely used in industry.

52. FLAME OR TORCH HARDENING

This method involves rapidly heating the surface of the steel above the critical point to the desired depth with an oxy-acetylene torch, or, more rarely, a flame obtained from another gas mixture, followed by cooling as desired.

For flame hardening of articles of cylindrical shape the part to be surface-hardened is fixed between centres of a lathe with oxy-acetylene torch and spray nozzle (intended for spraying water against the heated surface) held in the carriage.

Three main methods of flame hardening are used (Fig. 113). In the first method (*A*), the spindle of a lathe slowly rotates at a peripheral speed of about 2 mm per second, while an oxy-acetylene torch equipped with a head of equal width to that of the whole area to be hardened is employed for successive heating of the surface, followed by its immediate cooling. The whole surface area is hardened with only one revolution of the spindle. The main drawback of this method is the formation of a soft band several

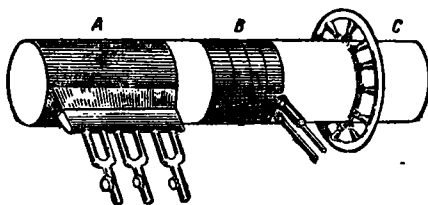


Fig. 113. Sketch illustrating three methods of flame surface hardening of a shaft.

millimetres wide at the junction between the two bands hardened first and last. Of course, it is possible first to gradually heat all the area to be hardened by rapidly rotating the spindle, and only then to cool it quickly. It is obvious that in this case the consumption of oxygen and acetylene is considerably increased, but a self-tempered band will not be formed.

The second method (*B*) involves successively heating and cooling separate areas of the surface to be hardened (the so-called zone heating). This method has the same disadvantage as the previous one, i. e., the formation of a soft, self-tempered, helical surface band.

In the third method (C), the surface to be hardened is rapidly heated with a multi-flame, annular torch-head, moving along the rapidly rotating shaft. An annular spraying device is used to cool the heated surface quickly. This continuous-successive flame hardening method, being the best of all three, gives no tempered bands, and a uniform hardness of the shaft treated results.

The acceptable depth of a hardened layer is from 5 to 10 mm. To lessen the case depth the surface layer has to be overheated. The thinner the layer to be heated, the more intensely the heat is conducted from the surface into the core of the article to be hardened.

In the continuous-successive flame hardening method the consumption of oxygen and acetylene can be calculated with formulae:

$$\text{Oxygen } w_o = 0.70 \sqrt{\delta} \text{ litre/sq cm;} \quad (19)$$

$$\text{Acetylene } w_a = 0.45 \sqrt{\delta} \text{ litre/sq cm;} \quad (20)$$

where δ —the assumed depth of the hardened layer in millimetres. When heating is simultaneous, the consumption of these gases increases by $1\frac{1}{2}$ times.

Torches are moved in the continuous-successive method of hardening at a rate determined with the formula:

$$v = \frac{72}{\delta^2} \text{ cm per minute.} \quad (21)$$

Duration of simultaneous heating equals:

$$r = 7\delta^2 \text{ seconds.} \quad (22)$$

The detailed heating process to be applied to individual articles is established by selecting the operation factors and checking not only the hardness and depth of a hardened layer, but its structure as well.

Apart from its great simplicity, the advantage of the flame hardening process is the possibility of its being applied to the selective hardening of large steel forgings and castings, which because of their size or shape are difficult or impossible to treat by any other method, including the most perfect one, i e., surface hardening with heating by electrical induction. Therefore flame hardening is fairly extensively used in the machine-building industry at the present time.

There are, however, serious drawbacks in the flame hardening process, the most important of which is the overheating of the hardened layer. Even when the process is adequately adapted to the specified conditions, some overheating cannot be avoided in practice. Another disadvantage is the difficulty of adjusting the depth of the hardened layer.

53. SURFACE HARDENING IN AN ELECTROLYTIC BATH

Heating parts in an *electrolytic bath* is based upon a physical phenomenon called the "cathode effect". The basis of this is the fact that if a direct current at high voltage (from 200 to 220 V) and considerable density (from 3 to 4A/sq cm) is passed through an electrolytic bath of solution of sodium carbonate (5 to 10 per cent) or potassium carbonate (potash), a thin layer of hydrogen bubbles is formed on the cathode (Fig. 114). Because of the low conductivity of the hydrogen resistance to the passage of an electric current is greatly increased, and the cathode is heated up to a very high temperature (about 2000° C).

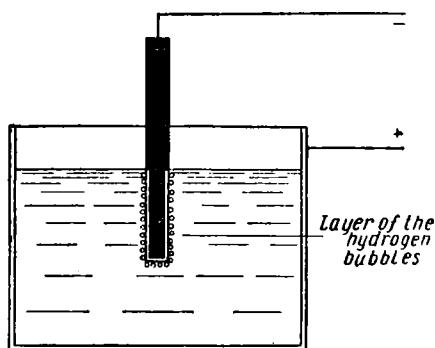


Fig. 114. Sketch illustrating the hardening operation in an electrolytic bath.

in the electrolytic bath, becomes a cathode. The piece may be immersed in the bath at the desired depth, depending on the part that is to be hardened.

Heating is conducted very rapidly, being completed in several seconds. As soon as the piece has been heated the electric current is switched off and the article is allowed to cool in the electrolytic bath, the latter being hardly heated at all by the current passed through it. Hence the electrolytic bath may be said to serve simultaneously as a heating furnace and a quenching tank.

The whole process of heating articles in the electrolytic bath, followed by quenching in it, may be completely automatized. This makes this method of surface hardening very suitable for application in mass production.

54. INDUCTION SURFACE HARDENING

Of all methods of surface hardening the best is an operation involving rapidly heating the surface of the steel to the desired depth by means of electrical induction and subsequent cooling. It has all the advantages of surface hardening—a very high productivity, and pos-

sibilities of total automatization—and produces a high-quality layer. It was worked out by V. P. Vologdin.

Heating steel by means of the high-frequency electric current is a particular case of induction heating. As is well known from physics, the true nature of the induction heating method consists in that when the metallic article is placed in a varying magnetic field a so-called eddy current (or Foucault current) is induced in it. This eddy current produced within the metal itself by induction is employed to heat it. The eddy current is always generated in the metallic article when it is placed within a varying magnetic field. In many cases, e. g., in transformers, the eddy current is harmful, as a part of the electric power transformed is consumed in a completely useless heating of the transformer itself. Special measures are taken, therefore, to reduce harmful eddy current to a possible minimum level. In induction heating, however, eddy current is useful, for it is precisely

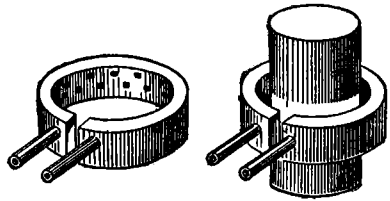


Fig. 115. Sketch illustrating the induction heating of articles to be hardened.

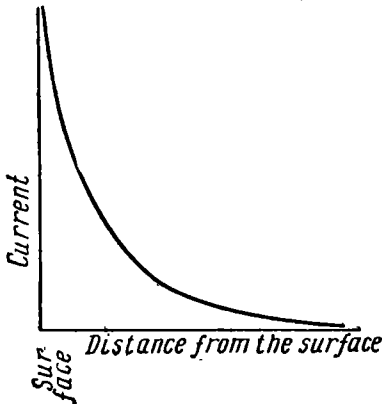


Fig. 116. Diagram showing the cross-sectional distribution of eddy currents taking place in heating articles by means of high-frequency electric current.

this that is employed to produce heat.

Cylindrical parts are heated by induction by being placed within a copper induction coil, or copper inductor (Fig. 115), through which an alternating current is passed. In the medium surrounding the inductor a varying magnetic field is produced, which penetrates the article and induces an eddy current in it. A peculiar feature is the uneven distribution of the induced eddy current in the cross-section of the article, as shown in Fig. 116, with the current density being at its greatest at the surface and diminishing rapidly with distance from it. This phenomenon has been termed an electromagnetic

ic surface effect or, briefly, a skin effect. The higher the frequency of a varying magnetic field, i. e., the higher the frequency of an alternating current, the more uneven will be the distribution of the eddy currents in the cross-section of the object being heated.

The fact that the eddy current density is greater at the surface of the article heated than in the centre accounts for its non-uniform heating, with the result that its surface layer is heated very rapidly to high temperatures, while the central portion remains cold or is heated only very slightly (through thermal conductivity only). Such is the manner of heating by high-frequency electric current.

High-frequency motor generators and valve generators, or valve oscillators, are used as sources of power supply. The motor-generator sets are employed in cases where the frequencies required do not exceed 10,000 c.p.s., and valve generators in those cases where higher frequencies are needed.

Surface heating by high-frequency electric current is determined on the basis of the following main factors: 1) frequency, and 2) time of heating.

Practice has shown that if the depth of hardened layer required is designated by a (cm.), the frequency f should satisfy the inequality:

$$\frac{150}{a^2} < f < \frac{2,500}{a^2} \quad (23)$$

whereas the optimum frequency, the equality:

$$f_{\text{opt}} \approx \frac{600}{a^2}. \quad (24)$$

Table 19, compiled according to data given by A. D. Demichev and S. V. Shashkin, shows recommended frequencies for the surface heating of steel articles according to the depth of hardened layer required.

The depth of hardened layer to be obtained in surface heating by this method is selected according to various working conditions of the object.

For parts subjected to wear in service the depth of a hardened layer ranging from 1.5 to 2 mm will in most cases be sufficient; and for parts whose surface should be hardened to withstand crushing or squirting this depth is increased to from 4 to 8 mm. For rolls used in cold-rolling depths of 10 mm and over are suitable.

The figures given in Table 19 may be used provided that the depth of the hardened layer is considerably less than the diameter of the article heated, otherwise the heating conditions will be drastically deteriorated. If a depth a of the hardened layer is less than or equal to 0.1 of diameter D (cm), the minimum frequency of alternating current should then be:

$$f_{\text{min}} \geq \frac{250,000}{D^2} \quad (25)$$

but if $a = 0.3D$,

$$\text{then } f_{\text{min}} \geq \frac{30,000}{D^2}. \quad (26)$$

Table 19

**Recommended Frequencies To Be Used in Surface Heating
of Steel Articles**

Specified depth of hardened layer, mm	Frequency in cycles per second (c.p.s.)			Generators recommended
	Maximum	Minimum	Optimum	
1.0	250,000	15,000	60,000	Valve generator
1.5	100,000	7,000	25,000	Valve generator, or motor-generator
2.0	60,000	4,000	15,000	Motor-generator, with frequency of 8,000 c.p.s.
3.0	30,000	1,500	7,000	Motor-generator, with frequency of 8,000 c.p.s.
4.0	15,000	1,000	4,000	Motor-generator, with frequency of 2,500 c.p.s.
6.0	8,000	500	1,500	Motor-generator, with frequencies ranging from 2,500 to 1,000 c.p.s.
10.0	2,500	150	500	Motor-generator, with frequency of 500 c.p.s.

Note: The frequencies indicated for motor-generators are those produced by them.

The heating time may be determined by calculation, but the computing formula is rather complicated. Therefore in practice the following procedure is usually adopted: When the frequency has been evaluated (e. g., by using Table 19) and some arbitrary period of time has been selected, the generator is adjusted to the determined frequency and the time relay to the chosen time. The test hardening is then performed. When this has been completed some method of examination, e. g., metallographic, is used to determine the depth of hardened layer actually obtained. On the basis of how near the obtained depth of hardened layer approaches the specified value, the surface heating factors are adjusted. In first test hardening operations the heating time may be a few seconds (usually 5 to 10).

Surface hardening by applying high-frequency electric current may be accomplished in various manners. The principal ones will be discussed below.

In simultaneous hardening the part to be hardened is rotated within the inductor (Fig. 117), the width of which is equal to the breadth of the surface area to be hardened. The part should be rotated to ensure uniform heating, as the symmetry of a magnetic field is broken at the place where the busbars are connected to the ring of the inductor. When the time of heating given for the process has elapsed and the time relay has turned off the inductor from the generator,

the object under treatment is immediately cooled by water spraying. In this operation another relay interlocked with the first one turns on the water supply system, and a great number of water jets are directed through holes in the interior surface of the inductor against the surface of the article.

In the continuous-successive method of hardening the part to be hardened performs two movements simultaneously; as it rotates it moves along the axis (Fig. 118), whereas the inductor remains stationary (in some methods the reverse is true—the article is stationary and the inductor moves along it). In such cases the heat is applied progressively,

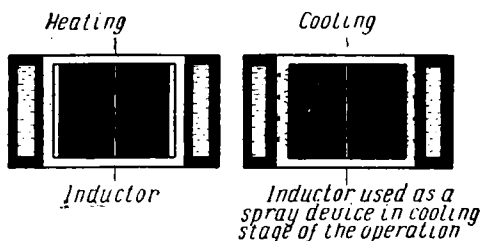


Fig. 117. Sketch illustrating the method of simultaneous surface hardening by means of high-frequency electric current.

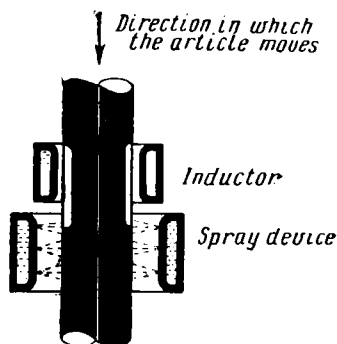


Fig. 118. Sketch illustrating the method of continuous-successive hardening by means of high-frequency electric current.

while cooling is effected by means of a separate spraying device mounted at some distance from the inductor.

In surface hardening of articles of flat shape the so-called loop-type and zigzag-type inductors are employed (Fig. 119). The inductor is made to perform a reciprocating motion in order to equalise the heat distribution and ensure uniform heating.

Gears of small diameter are heated in an annular inductor, similar in form to that of a gear. The teeth of large diameter gears (with a module exceeding 8) are heated individually, i. e., one at a time, by means of the loop-type inductor.

At the present time surface hardening by high-frequency electric current is applied to a wide variety of articles, e. g., crankshafts, gears, cams, machine-tool guides, railheads, ploughshares, hydraulic turbine blades, caterpillar track eye rings, cylinder liners for diesel engines, rolls, and a great many other machine parts.

The uses of this method are limited only by the complexity of the inductors which have to be manufactured, especially for treating parts of complicated shape. Induction hardening is therefore most common-

ly used in a continuous mass-production process, applied to articles of the same type.

Now let us consider the peculiar properties of a surface-hardened layer and a part subjected to induction hardening.

Unlike most other methods, this one does not as a rule cause any overheating in the surface layer, the structure of which is composed of finely acicular (or structureless) martensite. The hardness of a hardened surface layer is always somewhat *higher* (by 1 to 2 numbers on

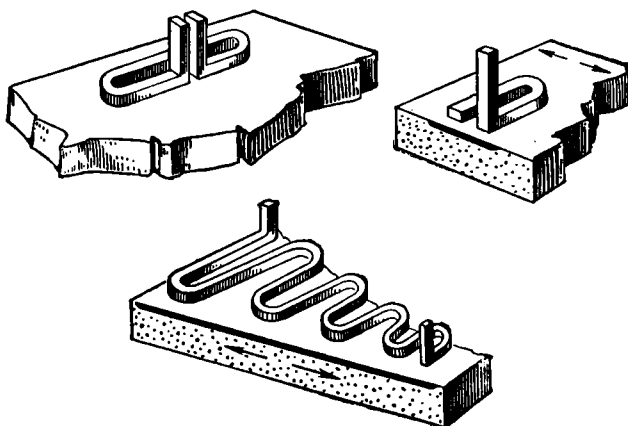


Fig. 119. Sketches illustrating the loop-type and zigzag-type inductors to be used in surface hardening of flat shape articles by means of high-frequency electric current.

the Rockwell "C" scale) than that of the surface layer of an article made from the same grade of steel and subjected to an ordinary hardening. Consequently, the wear resistance of this layer will also be higher. This fact is first of the considerable advantages, from the standpoint of metallography, inherent in articles hardened by applying high-frequency currents.

Investigations have shown that *mechanical properties*, too, especially such an important one as *fatigue strength* or endurance limit, are *higher* in articles with a surface layer hardened by high-frequency electric current. This is explained by the fact that in this treatment compressional stresses are set up within the article. We know that the major danger to articles in service is that of the tension stresses set up in it, when under the conditions of alternating bending; these are partially suppressed by the internal compressional stresses produced in hardening by induction.

Chapter X

CASE-HARDENING OF STEEL

55. ESSENCE AND CLASSIFICATION OF TYPES OF STEEL CASE-HARDENING PROCESSES

In the process of steel making the chemical composition of steel, particularly its percentage of alloying elements, is subject to changes. Carbon and alloying elements dissolve and are uniformly distributed in the liquid steel. In the whole bulk of a steel ingot and of the sheets, rods and pipes rolled from this ingot there is also a comparatively uniform distribution of the carbon and alloying elements (if the non-uniformity in composition on account of segregation is disregarded).

There is, however, another method of changing the chemical composition of steel, *the case-hardening treatment*. The changes obtained by this method are characterised by two features, i. e.: 1) the chemical composition is changed when the steel is in a solid condition; 2) the change occurs only in the surface layer of the article treated.

But, to what extent the change in chemical composition of only surface layers may be considered as satisfactory? Obviously this is satisfactory only in those cases when it is necessary to change the properties of the surface of an article, making them different from those of the interior portion, or where a change in the chemical composition and mechanical properties of the surface layer alone is sufficient, while the composition and properties of the interior portion are more or less irrelevant. Let us consider actual examples of each of these cases.

There are a great many articles for which it is of vital importance that the *surface hardness* should be as *high* as possible and the *wear-resistance* as *great* as practicable (since they are subjected to wear in service), and that they should possess at the same time a fairly *high impact strength*. Such articles include, for example, some kinds of gears, cams, eccentrics, ratchet wheels, etc. If high-carbon steel were used to manufacture such articles they could easily be given a high hardness by some appropriate heat-treating operation, but their impact strength would simultaneously be considerably decreased. On the other hand, if these articles were made of a low-carbon steel, a high impact strength would be combined with a low hardness. If the carbon content of the surface layers of articles made from a low-carbon steel is increased, then obviously the problem is solved, i. e., the hardness of the surface layers will be drastically increased by

hardening, while the high impact strength peculiar to the low-carbon steel will be retained, as it changes very little in hardening. This is exactly the procedure adopted in carburising.

Sometimes, however, parts are also carburised that are not required to possess a high impact strength, e. g., templates or gauges. These articles could be manufactured of a high-carbon steel (as in fact is largely done in practice). Instead they are sometimes made from a low-carbon steel and then subjected to carburising. In some cases this is due to the fact that a high-carbon steel is considerably more expensive than a low-carbon one, and a carburised template made of a low-carbon steel is therefore cheaper than one made of a high-carbon steel. In other cases such articles are carburised because the plant lacks high-carbon rolled stock of suitable shapes and dimensions, whereas at any machine-building plant there is ordinarily available a very ample variety of rolled shapes and sections of low-carbon steel.

Parts that are required to have a *high corrosion resistance* or *high resistance to scaling* at elevated temperatures may be manufactured either of acid-resistant or non-scaling alloy steels, or of a plain low-carbon steel, which is then subjected to an appropriate case-hardening (chemical-thermal) treatment to increase its resistance to corrosion or scaling. In most cases the results obtained are almost identical, for both corrosion and oxidation start at the surface of the article. The question now arises as to why a non-scaling alloy steel should be used if its resistance to scaling at elevated temperatures is equalled by that of a plain carbon steel subjected to a case-hardening treatment, while the former is certainly more expensive than the latter. In answer, it should be borne in mind that the surface, non-scaling layer is comparatively thin, and if it is destroyed in erecting (e. g., in bolting together or welding) even in only one place, then the total effect of a case-hardening treatment will be lost. The designer has therefore to take a separate decision for each part as to whether it should be manufactured from an alloy steel or from a plain carbon steel subjected to a subsequent case-hardening treatment.

The case-hardening treatment is applied to articles required to possess an increased wear-resistance, corrosion resistance and resistance to scaling in their surface layers. To increase the surface hardness and wear-resistance of machine parts the machine-building industry employs carburising, nitriding, cyaniding and chrome-plating. To increase the corrosion resistance articles are subjected to nitriding, chrome-plating and siliconising (in which a steel is caused to absorb silicon), and to improve resistance to scaling calorising, in which the steel is made to absorb aluminium, is carried out. Methods have also been developed for increasing the resistance

of a steel to scaling by causing it to absorb boron ("borisation") or beryllium ("beryllisation").

The case-hardening treatment should not be used in cases where changes in properties, such as tensile strength, heat resistance and magnetic properties, extending throughout the whole bulk of articles, are required.

56. FUNDAMENTALS OF CASE-HARDENING PROCESSES

Case-hardening of steel consists in saturating the surface of a hard steel with carbon, nitrogen and certain alloying elements. This operation is performed as follows: solid, liquid, or gaseous substances containing the element which is to be made to saturate the surface layer of a steel are brought into close contact with a steel article. Under such conditions the steel is caused to absorb these elements and thus change its composition at the surface.

If a certain element combines with iron to form a chemical composition, then the thin surface layer of a steel part will consist of this composition. Further penetration of the element into deeper layers of the article depends upon whether this element is capable of forming a solid solution with iron. In this case only can the element diffuse into the metal.

All the elements that are used in the industry at the present time to saturate the steel undergoing the case-hardening treatment, i. e., carbon, nitrogen, chromium, silicon and aluminium, are capable of forming solid solutions. Therefore theoretically all these elements can be made to diffuse in the steel at any temperature, but in practice, the steel is caused to absorb these elements always at elevated temperatures only. This is explained by the following three reasons.

First, the solubility of all the above-listed elements in iron increases with a rise of temperature. This applies especially to carbon. Its maximum solubility in alpha iron is about 0.02 per cent at 723°C, whereas at the same temperature the gamma iron can dissolve about 0.8 per cent carbon, and with rising temperature the carbon solubility reaches a value of about 2 per cent (at a temperature of 1130°C).

Secondly, the diffusion rate markedly increases with rising temperature, as is clear from Fig. 120. For example, with a temperature rise from 925 to 1100°C the diffusion rate of carbon in iron increases by over 7 times; the diffusion rate of chromium in iron increases by over 50 times as the temperature rises from 1150 to 1300°C.

Finally, high temperatures in the case-hardening treatment are necessary for decomposing the substances surrounding the steel articles to be treated, the decomposition products of which form a source of active atoms of elements to be used for saturation of the sur-

face of a steel. These substances are carbonates, to be used in carburizing, ammonia, to be used in nitriding, and ammonium carbonate, to be used in calorising.

From what has been said about the temperature effect in the case-hardening treatment the conclusion may be drawn that the temperature should be increased. The main obstacle to this is the considerable drop in the service life of muffles, retorts and other parts of furnace equipment with increase in temperature; in electric furnaces the rise in temperature is accompanied by a decrease in the life of heating elements. In this respect, heating by applying the high-frequency currents is a very interesting and promising method.

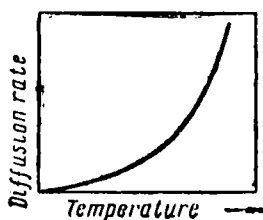


Fig. 120. Diagram showing the relationship between diffusion rate and temperature.

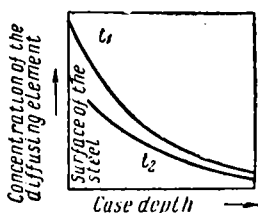


Fig. 121. Diagram showing the relationship between the change in concentration of a diffusing element across the section of a part to be treated and the temperature of the process ($t_1 > t_2$).

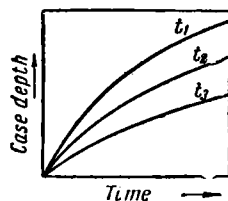


Fig. 122. Diagram illustrating the relation of time of case-hardening treatment to the depth of a diffusion layer (case) at various temperatures of the process ($t_1 > t_2 > t_3$).

It should be also borne in mind that the increasing of the temperature in the case-hardening processes adversely affects the steel structure. First of all, at very high temperatures the structure of steel in its central portion deteriorates. The case-hardening treatment at elevated temperatures may drastically worsen the structure of the diffusion layer itself. With rising temperature not only does the diffusion rate of an element in the steel increase, but the rate at which this element is transferred from the surrounding medium to the surface of an article increases still faster. In other words, the amount of an element deposited on the surface of the article is greater than that which can diffuse into the steel. As a result a very unequal distribution of the concentration of an element is obtained in the surface layer (Fig. 121), and a very high concentration of the element conducive to the formation of chemical compounds (carbides, nitrides, etc.) imparts a high brittleness to the surface layer, possibly even causing its scaling off. These factors limit the extent to which the temperatures of case-hardening processes may be increased. However,

the above-mentioned shortcomings of the high-temperature case-hardening treatment should not be feared, as they can be comparatively easily obviated in many cases.

Fig. 122 shows the effect of temperature upon the thickness of a diffusion layer. From this figure it is apparent that the thickness of a diffusion layer actually increases with time, but that this increase is not proportional to time. For example, while the thickness of such layer is a mm in one hour, it is less than $2a$ in two hours. Therefore, when it is said that the thickness of a diffusion layer increases by b mm in each hour, this refers to the mean values of increase in thickness usually obtained in practice.

57. CARBURISING

Carburising is the most widely used method of case-hardening. It consists of saturating the surface layer of a steel article with carbon.

Articles subjected to carburising are usually manufactured of steels of a relatively low-carbon content, such as 0.20 to 0.25 per cent. The steels ordinarily chosen for carburising include both plain carbon grades (Cr. 1, Cr. 2, Cr. 3, 10, 15, 20) and alloy steels (grades 15X, 20X, 12XH2, 12XH3, 18XHBA, etc.). The carburising process should be adjusted so as to obtain a carbon content in the surface layer of objects treated of 0.8 to 0.9 per cent, and in any case not exceeding 1 per cent. With a higher carbon content the surface layer or case becomes very brittle because a coarse cementite network is formed in it. The structure of a carburised case is shown in Fig. 123.

In modern practice carburising may be carried out with solid, liquid or gaseous carbonaceous mixtures. But in industry two methods of carburising are now mainly used, i. e., pack carburising

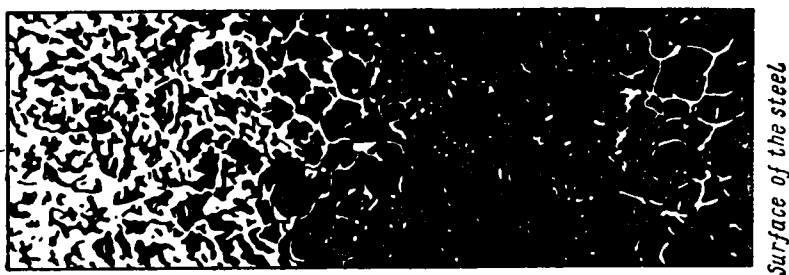


Fig. 123. Microstructure of a carburised zone or case. The surface of the part is on the right, 100 $\times$.

(with solid carburisers) and gas carburising (in contact with carbonaceous gases).

Pack carburising is performed as follows. Articles to be carburised are packed in close containers or "boxes", completely surrounded with solid carbonaceous materials, called *carburisers*. The packing operation requires care to ensure that sufficient compound is used to supply the carbon required, that flat surfaces of parts are separated by carburising compound and that no parts contact the sides of the pot or box (Fig. 124). The distance between parts to be carburised

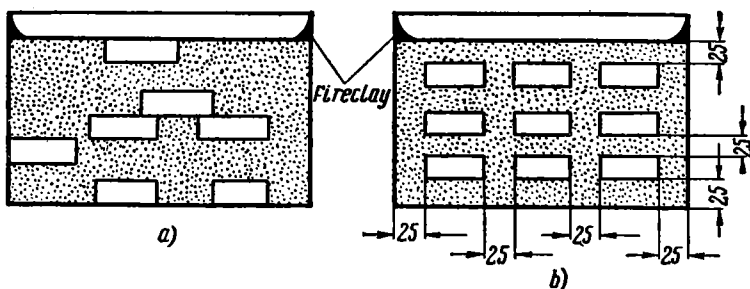
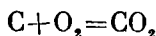


Fig. 124. Sketch illustrating methods of packing steel articles in the carburising box.

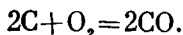
a—wrong; b—correct.

as packed should be not less than 20 to 25 mm, and that between the parts and the sides of the pot should be of the same order. With the packing completed and the carburising compound slightly tamped, the container is covered and all openings are luted with a non-cracking loam or clay in order to prevent the air from penetrating into the pot, or else the carburising compound will burn out in heating and the parts will be oxidised. The pots are then placed in the heating furnace and brought gradually to the carburising temperature of between 860 to 920°C. They are held at the carburising temperature for a period of several hours to give the desired case.

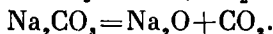
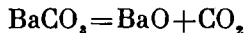
During heating in pack carburising two gases are formed in the pot, i. e., carbon monoxide (CO) and carbon dioxide (CO₂). These gases are obtained primarily as a result of the oxidation of the carbon forming the main portion of any solid carburiser, brought about by the oxygen from the air always available in the pores of the carburising compound. This process may be represented by the following equations:



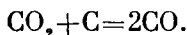
and



Other sources of carbon dioxide (CO_2) are the chemical compounds (barium, calcium or sodium carbonates) that are added to the carburising compound to act as energisers, being decomposed in heating:

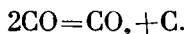


In the presence of the carburising material, e. g., charcoal equilibrium is established between these two gases, CO and CO_2 :



The higher the temperature, the greater the amount of carbon monoxide (CO) and the lower percentage of carbon dioxide (CO_2) the gaseous mixture will contain.

The main carburising reaction takes place at the surface of the steel parts:



After this reaction has produced the carbon in nascent form, it will, at carburising temperatures, be absorbed by the surface of metal and slowly migrate or diffuse towards the core, and at the same time the reaction between the carbon dioxide thus obtained and the charcoal will result in the formation of new molecules of carbon monoxide.

From this comparatively simple mechanism of the carburising process it is clear that in pack carburising, too, the carburising agents react with each other in gaseous form.

In the pack method the carburising agent must be in solid form, and practically every kind of solid carbonaceous material has been used or tried. The main component of all carburising compounds is coal, preferably charcoal, though bituminous coal is completely unsuitable for the purpose as it contains a considerable amount of sulphur. In some carburising compounds the charcoal is substituted by peat coal.

However, carburising compounds composed of only charcoal or peat coke alone produce a very weak carburising action. To increase it one or more chemical compounds are added to the carburising compound to act as energisers. The energiser may consist of barium, calcium or sodium carbonate (Na_2CO_3 or soda). Its function is to generate CO_2 , which reacts with carbon to increase the concentration of CO or to liberate nascent gases which may reduce the CO_2 . Fig. 125 illustrates sufficiently well the effect of barium carbonate

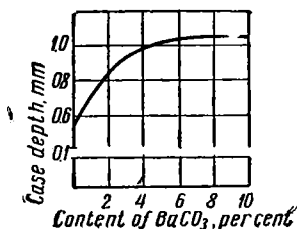


Fig. 125. Diagram showing the effect of the quantity of barium carbonate (BaCO_3) in the carburising compound upon the case depth as produced in carburising at 900°C for a period of 5 hours (G. F. Pulovsky).

(BaCO_3) added to the carburising compound. Table 20 gives the compositions of the carburising compounds most frequently used in practice.

Table 20

Compositions of the Solid Carburisers Most Frequently Used

Name of carburiser	Ingredients	Per cent by weight	Notes
"Bondug" carburiser	Charcoal (birch charcoal)	74 to 78	To be delivered in ready-made condition
	Barium carbonate	12 to 15	
	Sodium carbonate	1.0 to 1.5	
	Calcium carbonate	3 to 5	
	Fuel oil (or molasses)	4.5 to 5.0	
	Charcoal	90	To be prepared at the respective plants
	Sodium carbonate	10	
	Charcoal	90	Ditto
	Barium carbonate	10	
	Peat coke	85 to 90	Ditto
	Sodium carbonate	15 to 10	
Carburiser of the semi-coke type (U.S.S.R. State Standard type)	Grains of activated bituminous semi-coke, covered with a film of barium carbonate (the total weight of the latter amounts to 10-15 per cent)		To be delivered in ready-made condition

The active base of carburising compounds is usually charcoal or mixtures of charcoal and charred bone, which, with the coke, is generally crushed to pea size (from 3 to 8 mm), dried and screened (through screens with meshes 2 to 3 mm in size) to remove the dust. The screened charcoal is thoroughly mixed with barium carbonate or sodium carbonate and is stored in dry bins. The old carburising compound can be used again after being renewed or replenished by adding about one part of new compound to two parts of old compound. When old compound is used it is generally necessary to leave it exposed to the air for one to three days before it is used again to permit it to absorb carbon dioxide and moisture from the air, an action referred to as regeneration.

The carburising pots or boxes should be placed in a heating furnace maintained at a temperature of 700° C, and gradually heated up to the carburising temperature. In practice the time of heating is specified for each pot individually. Approximately it can be assumed that each 100 mm of width or height of the carburising pot requires about 2 hours in order to be heated.

The pots are held at the carburising temperature for a length of time depending upon the extent of the carburising action desired (for sufficient time to acquire the desired case depth). On an average it can be accepted that to produce a case depth of 0.1 mm the pots must be held at the carburising temperature for 1 hour. In practice, the following relationship has been established between the time of holding at a temperature of 920 to 930° C in pack carburising and the specified case depth:

Desired case depth, mm	Holding time of heating for carburising, hours
0.4 to 0.7	4 to 5
0.6 to 0.9	5.5 to 6.5
0.8 to 1.2	6.5 to 10
1.0 to 1.4	8 to 11.5
1.2 to 1.6	10 to 14
1.4 to 1.8	11.5 to 16
1.6 to 2.0	14 to 19
1.8 to 2.2	16 to 22
2.0 to 2.4	19 to 24

For the majority of machine parts the case depth is accepted as ranging from 0.5 to 2.0 mm.

Control of the case depth is performed as follows. Two or three holes are drilled in the carburising pot through which are inserted the test bars 5 to 6 mm in diameter and 200 to 300 mm long, made from the same steel as articles to be carburised. These test bars are also called "witnesses". Test bars are inserted into the carburising pot and packed with the work in such a manner that their greater part remains inside the pot, with only their ends projecting through the holes so that they can be easily withdrawn from the carburising pot without opening it. As soon as the time specified for the carburising process on the basis of computation has elapsed, one test bar is withdrawn from the carburising pot, quenched in water, and broken by hammer. The fractured surfaces of the test bar will show a coarse-grain structure surrounded with a dull-gray envelope (Fig. 126). This envelope, or rim zone, is the carburised layer, and its width is used in judging the case depth. The method of judging the depth of carburisation by fracture is very simple but is not very accurate. If greater precision of measurement is desired than can be obtained

by this method, the cross-section of the specimen may be polished and etched, and the case depth may be measured by a microscope. When the depth of carburisation as determined by the fracture of the test bar proves to be of the desired value, the carburising process is ended and the pots are removed from the furnace. If the case depth proves to be less than the specified value, the carburising operation is continued, and after some period of time the depth of carburisation is checked again by the fracture of the second test bar.

There are certain articles of which only certain parts are to be carburised, a process referred to as selective carburising. This operation consists of coating the parts of the surface not to be carburised with some special substances that prevent penetration of the carburising gases, e. g., a mixture composed of fire-clay, sand and asbestos, prepared with an admixture of water glass. Though this method of coating is very simple, it

is used only rarely in the present time, as it does not altogether prevent the carburising gases from penetrating to the parts not to be carburised. This is explained by the fact that such a coating cracks on heating, and the carburising gases penetrate through the cracks thus formed. At the present time, therefore, parts that have to retain softness are protected from carburising by copper plating, preferably by electroplating in which method a copper coating of about 20 microns thick is obtained, providing a satisfactory protection. To improve the shielding properties of a copper coating, it is recommended to cover it with a layer of water glass.

Pack carburising has many shortcomings both from the process and the operative standpoints. The overall carburising time is excessively long, and apart from the articles to be carburised, it is necessary in this method to heat the pots and carburising compound with their combined weight considerably exceeding the weight of the articles. Preparing the carburising compound and packing the parts to be carburised is a very dirty work, in which a great amount of dust is unavoidable.

Thus the interest taken in the more perfect *gas-carburising* process becomes quite obvious. This is carried out in muffle-furnaces. Inside the muffle the special atmosphere consisting of a mixture of gases is created. In gas carburising it is necessary to maintain a continuous fresh stream of carburising gases to the carburising retort, or muffle, as well as to remove the spent gases from the muffle contin-

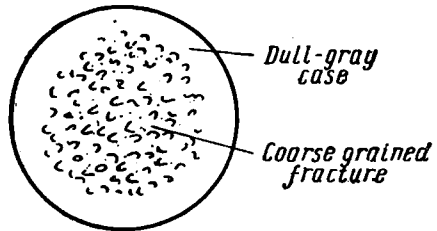
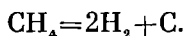
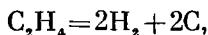


Fig. 126. Sketch illustrating the fracture surface of a test bar after carburising.

uously in order to obtain the correct mixture of gases inside it. The gaseous hydrocarbons most widely used in gas carburising are natural gas (methane), town or city gas, cracking gases (propane and butane), etc. All these carbon-bearing gases contain carbon monoxide and hydrocarbons that decompose in heating and deposit carbon upon the surface of articles to be carburised. For example, the mechanism of the carbon transfer may be represented as follows:



If the gas carburising is conducted with gases rich in hydrocarbons, they tend to break down and form soot carbon, which is deposited on the furnace walls and upon the work and interferes with the carburising process, since austenite does not absorb it at the temperatures ordinarily employed. Such carbon-rich gases are, therefore, to be partially burned prior to their use for carburising.

The amount of carbon deposited on the surface of the articles being carburised depends upon the stream of carburising gases conducted to the carburising retort or muffle, i. e., upon a rate of the supply of carburising gases. At a low rate of supply the carbon concentration in the case proves to be insufficient, i. e., too low, whereas at a high rate the most carbon is deposited upon the work to be carburised. Therefore in heating articles for carburising the rate at which the carburising gases are delivered into the muffle is adjusted low, reaching its maximum level at the first stage of the process and again slowing down in the final stage of operation in order to prevent the carbide network from forming in the carburised layer or case.

The rate at which the carburising gases are introduced into the muffle is adjusted with the aid of flowmeters and rheometers readings.

In gas carburising, the carburisation process proper takes only a slightly shorter period of time than in pack carburising. Yet the overall carburising time in the first case is shorter because there is no necessity to spend time on packing, unpacking and heating the carburising pots, as has to be done in the latter process.

In gas carburising the overall carburising time may be drastically reduced if the articles under treatment are heated by applying *high-frequency electric current*. Such a gas carburising unit is now in operation at the Likhachov Automobile Works in Moscow. The unit consists of a multi-coil cylindrical inductor 2 made from a copper pipe of square cross-section, installed vertically within a gas-tight shell 1 (Fig. 127). A stack of articles to be carburised 4, insulated from the inductor by ceramic bushings 3, is inserted within the inductor. Periodically a special device transfers the uppermost article into an annealing container for isothermal annealing. Simultaneously, a new article is

introduced from a prechamber below. Carburising is conducted at a temperature of 1050°C . To produce a case depth of 0.8 to 1.0 mm it is sufficient to hold the work at the carburising temperature for between 30 and 40 minutes. Investigations of microstructures have shown that in spite of such a high carburising temperature there is no substantial growth of the austenitic grains in steels of grades 18X1T and 30X1T, whereas in steels of grades 12X2H4A and 18XHBA a considerable growth of the austenite grains does take place.

After carburising, regardless of the agent employed in the process, the material is generally heat-treated. This is necessary for the two following reasons. First, if the steel of which the articles to be carburised are manufactured is originally coarse grained, then as a result of being held for a long time at elevated temperatures its internal portion, or core, will also acquire a coarse-grain structure and has to be refined. Secondly, in this case excess cementite often occurs as a network and induces brittleness, which is always to be avoided. Thirdly, the object of carburising is to produce as hard a surface resistant to wear and crushing forces as possible.

Therefore the heat-treating process as applied to carburised articles should normally consist of the following operations (Fig. 128, a):

1) normalising or hardening preliminary to carburising, conducted at a temperature of 850 to 900°C . This temperature corresponds to the A_1 temperature for the original steel containing about 0.2 per cent carbon, and to the same temperature for the carburised case with a carbon content of about 1 per cent;

2) quenching from a temperature somewhat exceeding the A_1 point, i. e., for a plain carbon steel from 760 to 780°C . In quenching from such temperature the outer zone will be fully hardened, with a maximum hardness resulting (60 to 64 Rockwell C), while the core remains very little hardened;

3) tempering at a temperature of between 150 to 170°C .

However, at many plants the heat-treatment of carburised articles is actually performed not in full, as indicated above, but with the normalising treatment omitted, i. e., the heat-treating process starts

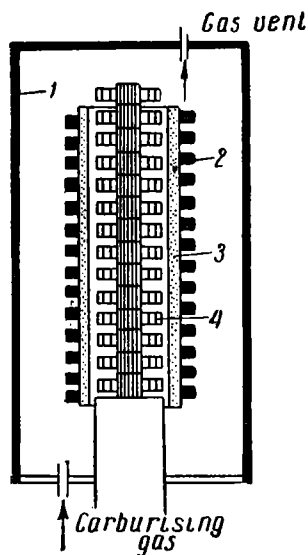


Fig. 127. Sketch illustrating the gas carburising unit in which the heating is effected by applying high-frequency electric current.

with quenching. There are two possible methods of such simplified heat-treating: 1) cooling the articles after carburising and then again heating them for quenching (Fig. 128, b); 2) quenching from the carburising temperature without cooling, i. e., the articles are withdrawn from the carburising box or from a furnace with a gas atmosphere and plunged into the quenching bath (Fig. 128, c).

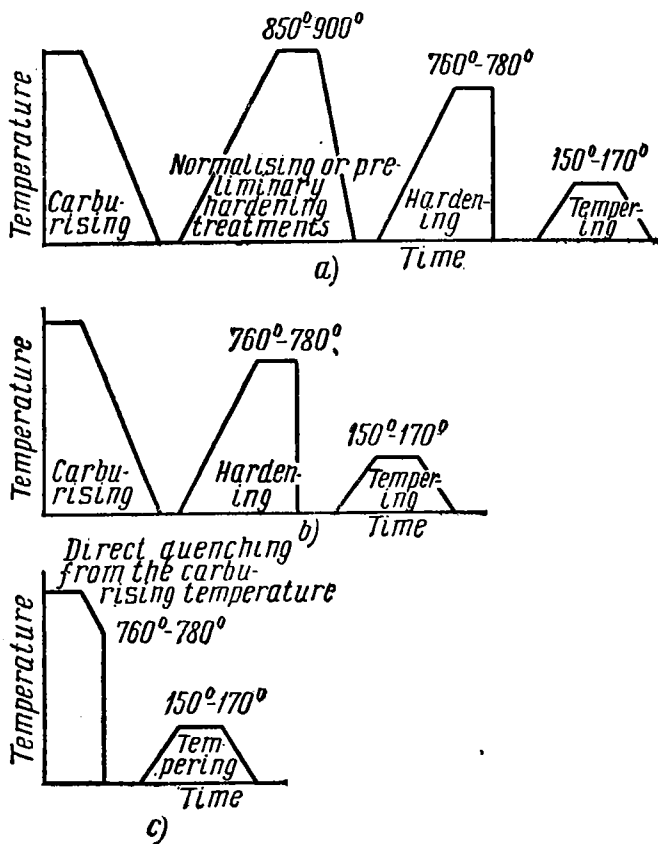


Fig. 128. Diagram illustrating several heat-treatments for carburised steel articles.

The advantages of both these methods of heat-treating carburised articles, especially of the second one, are obvious: the length of time required for the whole heat-treating procedure is curtailed, with the warpage, deformation, oxidation and decarburisation of the parts substantially reduced. These methods may be considered suitable in cases where the articles under treatment are required only to possess

a hard surface resistant to wear, i. e., where they are not heavily stressed in service, and where the mechanical properties acquired by the core in the heat-treatment are irrelevant and the cementite network causes no apprehension (e. g., some kinds of gauges, cams, etc). These methods may be suitable also in cases where the parts are manufactured of originally fine-grain steel, and where the formation of the cementite network is avoided by applying a carefully elaborated carburising procedure. In such a case there is no necessity to conduct normalising or quenching treatments preliminary to carburising. Such production conditions are created at automobile works, where the heat-treating practice is generally on a high technological level, but in most cases normalising and preliminary quenching treatments are to be considered as indispensable operations in the heat-treating process applied to carburised articles.

When articles of alloy steels are quenched from the carburising temperature, there is an excessive amount of residual austenite in the outer layer, decreasing the hardness of the case. To reduce the amount of residual austenite in the outer zone the carburised articles are quenched from a temperature of 750 to 800°C, after having been a little pre-cooled in the open air. The amount of residual austenite in the case produced in the quenching operation may be further diminished by the sub-zero treatment.

Defects in carburising most frequently to be met with are as follows:

- 1) case depth produced not complying with specifications;
- 2) distribution of carbon in high steps from the surface to the core;
- 3) hardness being too low;
- 4) non-uniform hardness (soft spots).

Wrongly prescribed or inaccurately controlled time of the carburising process is the main cause of the case depth's not conforming to specifications. When the case depth is too shallow, it can be corrected by repeating the carburising operation, while too deep a case cannot be corrected.

Application of too high a temperature in carburising may result in the carbon being distributed in high steps from the surface to the core.

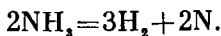
Using a weak or depleted carburising compound is the cause of too low a hardness of the case. This defect can be corrected by repeated carburising.

The presence of mill scale or rust and grease or oils on work not thoroughly cleaned preliminary to carburising most often results in the hardness of the case being non-uniform, with hard and soft places. Non-uniform hardness may be also produced if the carburising compound has been insufficiently tamped in packing, with some flat surfaces having been left uncovered.

58. NITRIDING

The process by which steel is caused to absorb nitrogen alone is called nitriding.

Nitriding is accomplished in muffle furnaces, usually pit-type furnaces. In the nitriding process the ammonia gas is introduced into a gas-tight furnace chamber heated up to a temperature of 500 to 600°C and decomposes according to the equation:



A very important fact to note is that the ammonia gas decomposes giving the nitrogen in nascent form, or *monatomic* nitrogen, which is the only form capable of entering the steel and diffusing in it.

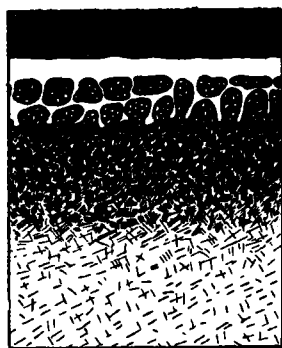


Fig. 129. Microstructure of a nitrided case (schematic diagram), 300X.

If the ammonia gas is decomposed outside the gas-tight retort of the furnace (in the so-called dissociating container), and then the dissociated ammonia is admitted into the retort, no nitriding of the steel will occur, because the *molecular* nitrogen contained in the decomposed ammonia (N_2) is not capable of diffusing into steel.

The structure of a nitrided case in plain carbon steels consists of a solid solution of nitrogen in ferrite (the so-called nitrogen ferrite) and the iron nitrides Fe_4N and Fe_2N (Fig. 129). In the structure of a nitrided case in alloy steels there are nitrides of the alloying elements, such as chromium, molybdenum and aluminium as well as the iron nitrides. In the outer zone of the steel the carbonitrides, chemical compounds of carbides and nitrides, are formed.

Nitrides produced by the combination of nitrogen with alloying elements possess a higher hardness, as compared with the iron nitrides. Because of this the nitriding process is primarily used to increase the *hardness and wear resistance* of the surface of the steel. Nitrogen ferrite, as well as nitrides, possesses a relatively high corrosion resistance. This fact determines the use of the nitriding process for protecting steel against corrosion.

Articles to be subjected to nitriding to increase their wear resistance are manufactured from steel of grade 38XM10A alloyed

with those elements (chromium, molybdenum and aluminium) that form nitrides which, firstly, possess high hardness and, secondly, contribute to a substantial increase in the hardness of the nitrogen ferrite, because they form finely dispersed particles.

The nitriding process applied to a steel to increase its wear resistance is a fairly long one. A period of about 10 hours at 500 to 540°C is usually required to produce a nitrided case 0.1 mm thick. Because of the extended period of time the minimum case depth has to be accepted: the accepted depth of a nitrided case does not usually exceed 0.5 mm.

The nitriding process can be controlled by adjusting the degree of dissociation of the ammonia gas, which, in turn, depends upon the rate of flow of ammonia introduced into the retort. The degree or percentage of dissociation is defined as the ratio of the volume of the decomposed ammonia in the gas mixture delivered into the furnace to the total volume of this mixture. The degree of dissociation can change within wide ranges (from 20 to 65 per cent).

To accelerate the nitriding process it is usually conducted in two stages: 1) holding at 500 to 520°C for a period of 8 to 12 hours at about 18 to 25 per cent dissociation of ammonia; 2) holding at 560 to 580°C at about 50 to 60 per cent dissociation of ammonia. The length of time required for the second stage of the nitriding process is determined by the case depth desired. The nitriding rate through the second stage of the process is 1.5 to 2 times faster than that of the first stage. After nitriding the work is allowed to cool slowly down to about 200°C in an atmosphere of ammonia at about 20 per cent dissociation.

The surface hardening produced by nitriding is really remarkable, varying from about 1,000 to 1,400 H_v, i. e., from 65 to 70 Rockwell C.

To prevent nitriding, parts to be protected are preferably tin plated. Though tin, of course, melts at the nitriding temperature, it is nevertheless retained on the surface of the work treated by the surface tension forces. At some plants local or partial protection against nitriding is obtained by using water or soluble glass as double covering for those parts of the work that are to be protected. After each coating with water glass the work is dried at a temperature of 100 to 120°C. However, the water glass coating is likely to deliquesce on the surface of the work during nitriding and cannot be used in those cases where an exact separation line between the nitrided and non-nitrided areas should be obtained.

The nitriding process is both more complex and expensive, as well as longer, than the carburising operation. Only one standard grade of alloy steel (38XM10A) can respond to nitriding for increased wear resistance. But the great advantages of nitriding are: first, it is accomplished at comparatively low temperatures and, secondly, quenching is not required after it. Because quenching is not involved,

the distortion of a part nitrided is negligible as compared with the warpage of the carburised article. Moreover, the nitrided surface retains its hardness at comparatively high temperatures better than any carburised steel.

Nitrided steels are therefore used for the manufacture of parts which require an extremely hard surface combined with a minimum distortion, such as cylinder sleeves and liners for Diesel engines, spindles for lathes, milling machines and grinding machines which work at speeds of over 2,000 r.p.m., etc.

Nitriding is the last process applied to an article before it is ground. Therefore the stock to be nitrided *must be finish-machined and fully heat-treated*. In most instances the steel is heat-treated to high mechanical properties before nitriding. For articles made of the 38XM10A grade steel the heat-treatment consists of quenching in oil from 940°C followed by drawing-back at a temperature of 600 to 650°C.

Articles of low-carbon plain steels are usually subjected to *nitriding for improvement of their corrosion resistance*. The nitrided case, as obtained from the steels of this series, has a high corrosion resistance. The case depth of 0.02 to 0.04 mm is sufficient to improve substantially the corrosion resistance of the steel in water or humid air. To produce the case depth of such thickness nitriding is performed at a temperature of 600 to 620°C for a period of 2 to 4 hours at about 40 to 50 per cent dissociation of the ammonia gas.

In great many instances nitriding to improve the corrosion resistance of steel may be substituted for nickel plating and chromium plating processes, being less expensive and easier to perform.

59. CYANIDING

Cyaniding involves heating steel in an environment that adds carbon and nitrogen simultaneously to the surface. This process has been named after the chemical compound of carbon with nitrogen, called *cyan* (CN). However, in the structure of the cyanided case there is, of course, no cyan at all, so there is no danger of poisoning.

Cyaniding is used for the same purpose as carburising and nitriding, i. e., for giving steel *high surface hardness and wear resistance*.

A considerable advantage of the cyaniding process, as compared with carburising and especially with nitriding, is its higher productive capacity; cyaniding takes several times less time than carburising.

Cyaniding is applied to parts made primarily of low-carbon carburising steels. These are usually parts of automobiles (limiter bolts, oil pump gears, drive worm screws, change-over switch shafts, steps, sleeves, brake cams, synchronising sleeves, speed-box gears,

etc.), motor-cycles (gears, shafts, pins, etc.), agricultural machinery (self-propelled combine gears), etc.

After cyaniding the steel is heat-treated in much the same way as after carburising.

There are two methods of cyaniding; *gas* cyaniding (carbonitriding), and *liquid* cyaniding (nitriding in a cyanide bath).

The *liquid cyaniding* process encompasses heating the steel in a fused cyanide bath, holding there for some period of time and then direct quenching from the salt bath. The quenching operation is usually performed by plunging the cyanided articles into oil.

Different types of cyanide baths of various compositions are used. The main, or activated, portion of any cyanide bath is composed of some cyanide salt, preferably sodium cyanide, NaCN , sometimes calcium cyanide, $\text{Ca}(\text{CN})_2$, potassium, ferrocyanide $\text{K}_4\text{Fe}(\text{CN})_6$, and others. The mixtures that are most frequently used range from 20 to 25 per cent sodium cyanide, but some contain as little as 5 to 7 per cent, and others a range of from 60 to 70 per cent. The base, or inerts, of the sodium cyanide baths may be various neutral salts, such as sodium carbonate, Na_2CO_3 , sodium chloride, NaCl , etc.

The carburising and nitriding action of a sodium cyanide bath takes place in steps or by a series of reactions which may be described as follows: first, the sodium cyanide is oxidised, with limited oxygen supply available from the air dissolved in the bath, then the cyanate, NaCNO , obtained in the first step breaks down, with the formation of *monatomic nitrogen* in nascent form and *carbon monoxide*, which in turn decomposes, giving carbon dioxide and carbon in the nascent state. Cyanide baths usually have a working temperature range from about 820 to 870°C. To produce a deep cyanided case (depth of up to 1.5 to 2.0 mm), the temperature of the cyanide bath is raised up to 900 to 940°C.

The time of holding in sodium cyanide baths at 820 to 840°C depends upon the desired depth of case to be cyanided, as follows:

Case depth, mm	Time of cyaniding, minutes
0.1	10
0.15	20
0.20	40
0.25	60
0.30	90
0.40	180

Usually the thickness of a cyanided case is accepted to be equal to 0.1 to 0.3 mm.

Since cyanide salts are very poisonous, care is required to avoid poisoning in handling them. The following safety rules must be complied in the cyaniding process:

1) cyanide baths must be provided with protective shells and effective exhaust ventilation;

2) rubber gauntlets should be necessarily used in weighing the cyanide salts, adding them into the salt bath, plunging the article to be cyanided into the cyanide bath, withdrawing it from the bath, and quenching it;

3) to prevent splashing of the fused cyanide solution, the articles, tongs, and suspension arms to be immersed in the cyanide bath must be completely dry;

4) eating and smoking are prohibited in the room where the cyaniding is being conducted;

5) at the end of the working hours, as well as before the dinner-break, the operator must thoroughly wash his hands with soap;

6) it is prohibited to take overalls outside the room where the cyaniding operation is being performed;

7) in working with cyanide baths it is prohibited to employ the same tongs and appliances that are used in working on the other furnaces, especially the saltpeter baths, for there will inevitably be an explosion if the saltpeter gets into the cyanide bath, or if, vice versa, the cyanide compound is brought into the saltpetrous bath;

8) articles heated in cyanide baths must be thoroughly neutralised and rinsed immediately after the quenching operation. Neutralisation is necessary to deprive the cyanide salts which may be retained on the articles of poison. Neutralising is done in tanks, in which a 3 to 5 per cent ferrous sulphate solution is provided. It takes from 5 to 10 minutes. After neutralising the articles are thoroughly rinsed in hot water at about 60 to 80°C for a period of 5 minutes.

Gas cyaniding (carbonitriding) is very similar to gas carburising and is performed in the same furnaces. In gas cyaniding a gaseous mixture of ammonia and some carburising gas, e. g., cracking gas, is used, and since this process involves heating the steel in an atmosphere that adds carbon and nitrogen to the surface, it can be said to consist of two simultaneous processes, *gas carburising* and *nitriding*. Therefore gas cyaniding is also called *carbonitriding*.

The gaseous mixture used in gas cyaniding is composed of 70 to 80 per cent (by volume) carburising gas and 30 to 20 per cent (by volume) ammonia gas. This composition is established and maintained during the cyaniding process by regulating the gas mixture by reference to readings of rheometers. Where the proportion of ammonia in the gaseous mixture exceeds 30 per cent, the process results in a very hard and brittle cyanided case. The temperature of gas cyaniding ranges from 820 to 850°C. The time of holding at the cyaniding temperature is computed on the following basis: to produce a case depth of 0.1 mm it is necessary to heat the steel at the cyaniding

temperature for a period of about 1 hour. So the rate of gas cyaniding is about 3 to 6 times slower than that of liquid cyaniding.

An advantage of the gas cyaniding process as compared with cyaniding in salt baths is that it does not involve the direct contact of operators with such poisonous matters as the cyanide solutions and is thus considerably less detrimental to the health of workers. Unlike cyaniding in salt baths, gas cyaniding may easily be made automatic. Gas cyaniding is best suited to mass and large-lot production, where it is required to treat by a standard technique large lots of uniformly shaped parts manufactured of the same steel grade and to be cyanided to the same case depth. But cyaniding in salt baths is a more flexible process, since it permits the simultaneous treating in the cyanide bath of articles of different designs that are made of various steel grades and are to be cyanided to different case depths. Therefore the cyaniding in salt baths is best suited to individual production.

The range of hardness for the cyanided case after quenching is generally 62 to 65 R_C .

60. DIFFUSION METALLISING

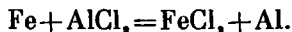
Apart from the above considered case-hardening processes—carburising, nitriding and cyaniding—other methods of saturating the surface of the steel with different metals (aluminium, chromium) and metalloids (e. g., silicon) are employed in the industry. All these methods are known as *diffusion metallising*.

Alitising (calorising). This process, involving the saturation of the surface layer of the steel articles with aluminium, is intended to increase its *resistance to scaling*. It is applied to parts working at elevated temperatures, such as fire-bars, crucibles, muffles, carburising pots or boxes, parts of gas producer units, economiser pipes, etc. Calorising is also performed on the blades of gas turbines in order to increase their resistance to wear by the particles of ashes and fuel which are incompletely burned.

Of the number of methods of calorising developed the technique of processing with *pulverulent mixtures* is now used in practice. This process is done as follows: The steel articles to be calorised are packed in containers, or boxes, with a suitable mixture consisting of: 1) aluminium powder, 49 per cent (by weight); 2) powder of aluminium oxide, finely crushed firebrick (chamotte) or burned fire-clay, 49 per cent (by weight); 3) ammonium chloride, NH_4Cl , 2 per cent (by weight). All these ingredients are thoroughly intermixed. The containers are then placed in a furnace and heated to 950 to 1050°C. They are held at this temperature for a period of from 3 to 15 hours,

depending upon the depth of case desired, which usually ranges from 0.3 to 0.5 mm.

On heating the ammonium chloride decomposes into ammonia and hydrogen chloride. These gases are made to displace the air from the container, and in reacting with the aluminium, the hydrogen chloride forms aluminium chloride, AlCl_3 , in a gaseous state. At the surface of the steel articles, an exchange reaction of the following type takes place:



The aluminium precipitated on the surface enters the steel and diffuses inwardly.

The base, or inerts (aluminium oxide, chamotte or fire-clay), is needed to prevent the aluminium powder from fusing into a liquid mass: only separate particles of aluminium melt and surround the particles of inerts, so that the porosity of a mixture remains preserved.

After calorising the steel is subjected to a diffusion annealing at 900 to 1050°C with the holding time at that temperature ranging from 4 to 5 hours. This heat-treatment needed to somewhat decrease the aluminium proportion in the outer zone of the calorised case is increased by 20 to 30 per cent.

The aluminium content in the calorised case may be as high as 40 to 50 per cent. The structure of the case (Fig. 130) consists of grains of a chemical compound, Fe_2Al_6 (outer zone), grains of a solid solution of aluminium in alpha iron pierced by needles of the same chemical compound (next layer), and grains of the solid solution not penetrated by needles of the chemical compound. Annealing causes the outer zone, composed only of grains of the chemical compound Fe_2Al_6 , either to disappear completely from the structure or to become considerably thinner.

Calorised articles are capable of withstanding temperatures up to 900°C without any oxidation of their surface. At higher temperatures oxidation does occur, but to a very slight degree. The diagram in Fig. 131 illustrates the high resistance of the calorised steel specimen to scaling at a temperature of 1000°C.

Chromium coating. The remarkable properties of chromium, such as high corrosion resistance, acid resistance, hardness and wear resistance, are used to advantage in chromium coating. In the machine-building industry, the method of coating the surface of the steel articles by electrodepositing the chromium layer on it is widely used. This method, known as *chromium plating*, is primarily employed to impart to the steel objects not only the resistance against corrosion but also a bright and pleasing appearance. However, the so-called "decorative" coatings of chromium applied by electrodeposition are insufficiently firmly bonded with the original steel base, to be relied upon under more exacting working conditions. Therefore the possibilities

offered by the diffusion of chromium into the steel seem to be very attractive.

One of the methods of *diffusion chromium coating* consists of packing the steel articles in containers or boxes with a special powder composed of 60 per cent chromium or low-carbon ferrochromium, 37 per cent aluminium oxide or fire-clay, and 3 per cent ammonium

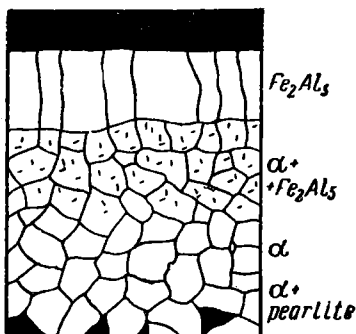


Fig. 130. Schematic diagram showing the structure of a (calorised) case, 300X.

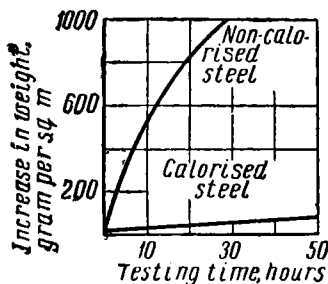
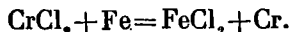


Fig. 131. Diagram illustrating the resistance to scaling of the Cr. 10 grade steel specimens, calorised and non-calorised, tested at a temperature of 1000°C (as given by A. N. Minkevich).

chloride. The containers are placed in a furnace, heated there to a temperature of 950 to 1050°C, and then held at that temperature for a period ranging from 8 to 15 hours. During this time the chromium enters the steel and diffuses inward, thus saturating a case of depth from 0.05 to 0.15 mm.

The reactions that take place in these boxes are very much similar to those that occur in calorising; first, a chromium dichloride (CrCl_2) is formed in a gaseous state and reacts with the iron at the surface of the steel objects as follows:



The chromium precipitated on the surface of the steel enters it and diffuses inward.

Methods of *gas chromium coating* are also developed. They involve passing a mixture of hydrogen and hydrogen chloride vapour through a furnace heated up to 1000°C, in which chromium or ferrochromium are placed. Reaction between these substances results in the formation of the gaseous chromium dichloride, which, in its turn, reacts with the iron according to the above reaction. Gas chromium coating, like all other processes of case-hardening, is a comparatively rapid process.

The chromium content in a case thus obtained is a very high one, ranging from about 25 per cent for low-carbon steels, to as high as

60 to 70 per cent for high-carbon steels. The structure of a case saturated with chromium consists of either grains of a solid solution of chromium in alpha iron or of grains of a complex carbide, $(Cr, Fe)_7C_3$. The surface hardening produced by chromium coating of high-carbon steels is really remarkable, the Vickers number reaching up to 1350.

Chromium coated articles of low-carbon steels are to be heat-treated when it is desired to increase the mechanical properties of their interior portion. The chromium coated tools should necessarily be heat-treated (hardened and tempered) to increase the strength and hardness of their core, or else the thin case might be pressed down or destroyed in service. The hardness of the case saturated with chromium does not change after heat-treatment.

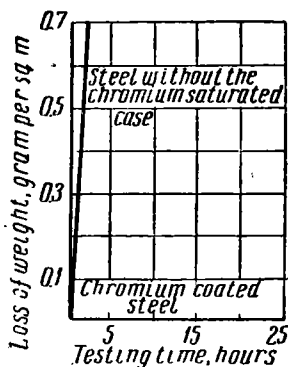


Fig. 132. Diagram illustrating the resistance to attack by 25 per cent solution of nitric acid of specimens made of Cr. 10 grade steel, chromium coated and without a chromium saturated case (as given by A. N. Minkevich and G. A. Faivilovich).

Fig. 132 illustrates the exceptionally high resistance of cases saturated with chromium to attack by nitric acid. Chromium saturated cases have a comparatively high resistance to attack by hydrochloric and sulphuric acids, but this cannot by far be compared with their remarkable resistance to attack by nitric acid. Therefore it is better to apply chromium coating to various parts of acid-proof apparatus and pipes rather than to manufacture such parts of stainless or acid-resisting steels.

In resistance to scaling chromium saturated cases are inferior to calorised cases. Therefore, it is not advised to employ the chromium coating methods for increasing the resistance to scaling.

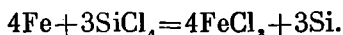
The high hardness and wear resistance peculiar to chromium saturated cases is made use of to increase the strength of dies, gauges and cutting tools. Superfine or dead smooth files with chromium saturated cases may be used even to file articles hardened to Rockwell C number 60 to 62. Experiments have shown that the life of bits provided with a chromium saturated case has increased by 1.5 times, the bits being manufactured of 12 grade steel.

Siliconising. Apart from chromium coating, the resistance of the metal to attack by acids may be increased by *siliconising*, that is, by saturation of the surface layer of steel or cast-iron objects with silicon.

Though there are methods of siliconising with the *pulverulent mixtures*, the *gas siliconising* method is generally preferred now.

This process consists of passing gaseous chlorine through a furnace heated up to 950 to 1050°C.

Chlorine reacts with the silicon carbide or ferrosilicium placed in a furnace along with the articles to be treated. This reaction results in the formation of gaseous silicon chloride, SiCl_4 , which, in turn, reacts with the iron at the surface of the steel. This exchange reaction may be represented as follows:



The reaction gives silicon in nascent form that enters the steel and diffuses inwardly. At the temperatures indicated above it takes about 2 to 4 hours to produce a siliconised case to a depth of 0.5 to 1.0 mm. The silicon content in the outer zone of the case is about 14 per cent. The structure of the siliconised case is composed of silicon ferrite.

Siliconising increases markedly the resistance of steel to attack by sea water, nitric, hydrochloric and sulphuric acids. According to data given by A. N. Minkevich, in testing a siliconised steel specimen in 10 per cent solution of boiling sulphuric acid for a period of time equal to 10 days its loss of weight was found to be only about 5 per cent, whereas a specimen made of the 1X18H9 grade acid resisting steel, tested under the same conditions, was fully dissolved by the acid (100 per cent by weight) in 30 hours.

The diffusion metallising processes include also: 1) *borisation*, or boronisation, involving saturation of the steel surface with boron; this process results in a very hard case ranging up to 1300 to 1400 Vickers number and an increased resistance to acids; 2) *beryllisation*—the saturation of the surface of the steel with beryllium; this results in an extremely hard case possessing a resistance to scaling nearly equalling that of a boronised case; 3) *molybdenising*, *tungstenising*, and *vanadising*; these processes consist of saturating the surface of steel with molybdenum, tungsten and vanadium respectively, and result in cases having to some degree a higher resistance to attack by acids.

Chapter XI

FINISHING OPERATIONS AFTER HEAT-TREATMENT

61. REMOVAL OF SCALE

Unless it is performed in a protective gas atmosphere, heating of steel is always accompanied by oxidation, i. e., formation of oxide or scale coating on the surface. The scale must be removed in cases where the rolled stock, billets, and articles are not to be machined but are required to be completely clean. This refers to ribbons, wire, sheet

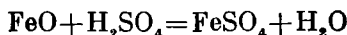
steel, and sheet blanks that are to be stamped, as well as to springs, annealed welded structures, etc. Scale remaining on rods and wires causes a severe wear of drawing dies. The life of stamping dies is also very much reduced by scale on sheets and sheet blanks. Scale on articles and welded structures to be coated with metal or with lacquer and paints is inadmissible; such articles or structures cannot be metal-coated, while lacquers and paints, though they can be applied to such surfaces, have a poor adhesion and scale off together with the scale. The scale should preferably be removed also from the surface of billets and objects that are subject to further machining, for though the scale is removed through machining, the cutting tools used in removing it undergo such a severe wear that their life is thus drastically reduced.

Thus in some cases it is imperative that the scale be removed from the surface of the steel, while in other cases it is desirable.

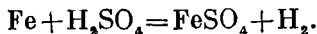
Two main methods may be employed for removing the scale: 1) *acid pickling* and 2) *abrasive blasting, involving the use of sand, shot or grit*. The metal-cleaning method may be chosen arbitrarily, depending mainly upon the quantity of work involved. The quantity of work involved in the acid pickling process is considerably less than that in other metal-cleaning methods. Therefore at metallurgical plants all the rolled stock to be cleaned (e. g., rods, wire, tubes, sheet metal) is subjected to acid pickling.

Acid pickling of metals is generally conducted in a 8 to 12 per cent (by volume) solution of sulphuric acid bath. A 20 per cent (by volume) solution of hydrochloric acid may also be used for descaling, but pickling in a hydrochloric bath is more expensive.

Acid pickling to remove the scale from a steel surface consists of two processes: 1) direct dissolution of the scale in an acid bath according to the reaction:



and 2) mechanical exfoliation of the scale by penetration of acid to the metal beneath, the reaction forming ferrous sulphate and hydrogen:



To lessen the attack on the base metal and reduce the danger of hydrogen embrittlement, the so-called inhibitors are added to the pickling bath. The inhibitors are usually substances of organic origin, such as serum preparation, compound of the hemp oil cake, preparations of sugar factory waste and oil refining products, etc. An inhibitor added to the pickling bath slows down or prevents the action of the acid upon the base metal by forming a protective adsorption film on its surface. Pickling is conducted at 40 to 50°C. The pickling

bath is heated to that temperature by passing steam through it. As the pickling bath is weakened in its action (being depleted of the sulphuric acid), its temperature is increased to 70 to 80°C. The acid concentration must not decrease below 3 to 4 per cent.

In pickling the steel may absorb hydrogen. This very dangerous phenomenon, causing a drastic drop in the ductile properties of the steel, is known as hydrogen embrittlement. It may be prevented either by special heating at 120 to 150°C for a period of 1½ to 3 hours or by allowing the steel to season at room temperature for not less than 2 days.

The rolled stock, billets, and articles removed from the pickling tank are thoroughly rinsed in streams of fresh water for several minutes, neutralised for 5 to 10 minutes to remove remainder of the acid by immersion in a weak alkaline bath, such as lime water or 1½ per cent (by volume) solution of sodium hydroxide (NaOH), and finally rinsed again in a warm water at 60 to 70°C to accelerate the drying operation.

Sandblasting for cleaning steel consists of directing jets of compressed air carrying sand composed of dry, sharp quartz grains against the surface of articles to be cleaned. The grain size of sand should be from 1 to 2 mm. Sandblast cleaning of exterior surfaces is quite satisfactory, but interior surfaces are cleaned rather less well.

A great and unavoidable disadvantage of sandblast cleaning is the evolution of a large amount of finest quartz dust, which seriously affects the health of workers. If a large amount of this dust is inhaled by a worker it causes a professional disease of the lungs, called silicosis. Therefore sandblast cleaning must be done in special rooms, the so-called sandblasting chambers, provided with an effective exhaust ventilation. The operators in charge of the sandblast cleaning, who are exposed to the dust during the operation, must use respirators and sandblast hoods to prevent the inhalation of silicate or quartz dust.

However, these means of protection are not completely effective, and exposure to the dust often caused tuberculosis. So at the present time the more sanitary *shotblast cleaning* method is being substituted for sandblasting. In this method, hard metallic particles or iron shot (often also called "steel grit") made of white iron, with a particle size of 0.5 to 2 mm, are blown at high velocity against the surface to be cleaned. The shotblasting is effected with the same equipment as is used in sandblast cleaning. With shotblast cleaning there is incomparably less dust than in sandblasting. A still more advanced method is the so-called *airless abrasive cleaning*. A type of machine employing the centrifugal method instead of pneumatic pressure for abrasive cleaning has been developed. This type of machine has a rotating head with cups or vanes, which takes small quantities of abrasive

material such as iron shot or grit. As each vane rotates it delivers the abrasive material at high velocity through slots in a shell and within a predetermined area on the surface of articles. The abrasive action of the material not only cleans the scale from article surface, but produces the cold hardening of the latter. In some cases this cold hardening of the surface becomes a special application of the shot-blasting and airless abrasive cleaning processes, which not only cleans but increases the fatigue life of articles. Cold hardening throws the surface layers under compression, and the compressional stresses contribute to increasing the fatigue life of the steel parts. This explains why the sandblasting and airless abrasive cleaning methods are used not only to remove the scale from the surface of the steel but as a means for its surface hardening.

62. ALKALINE-DETERGENT CLEANING AND DEGREASING

When articles and tools are heated in salt baths (saltpeter baths included), the salt adheres to the surface of the metal. On quenching in water not all the salt is dissolved; a part of it remains on the sur-

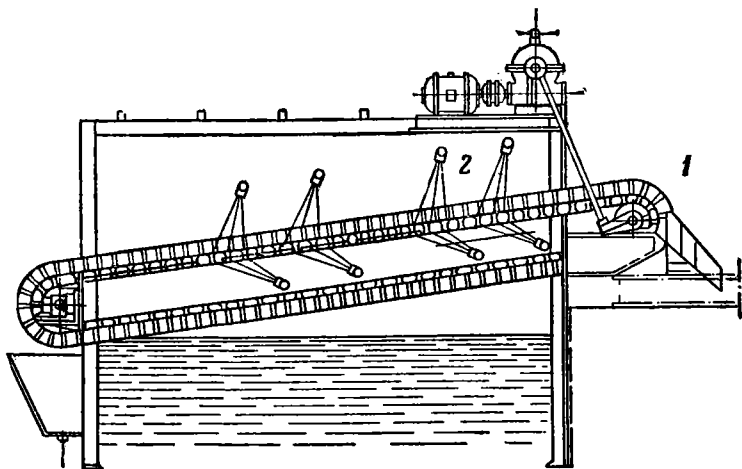


Fig. 133. Washing machine.

face, especially in corners, between the teeth of gears, etc. On quenching in oil, the adhering salt remains intact. To remove the remnants of salts in such cases the articles and tools have to be "soaked" in the boiling water. Articles and tools tempered in oil baths as well as those quenched in oil have to be degreased. Sometimes it is consid-

ered superfluous to clean the articles of oil, since it is said to be burned off in tempering. This is quite wrong. It is true that the oil actually burns off in tempering, but if the tempering is conducted not in flame furnaces but in electric furnaces, there is such a great evolution of obnoxious fumes in the shop that no exhaust ventilation can succeed in cleaning the air. Moreover, the soot deposited on the electric heating elements shortens their service life: it will be remembered that carbon is highly detrimental to steels and alloys of high electrical resistivity.

The articles and tools to be cleaned from oil are rinsed in a 10 per cent (by volume) solution of soda ash (Na_2CO_3) at 80 to 90°C, or in a 3 per cent (by volume) solution of caustic soda or sodium hydroxide (NaOH). Very often the salts and oil are removed from the work surface in one operation, the so-called "soak cleaning" or "vat cleaning", consisting in immersing the object in a tank of boiling soda solution and then rinsing.

In large scale production the degreasing operation is mechanised by using special washing machines (Fig. 133). In this machine the parts to be cleaned are slowly moved on conveyer belt 1 through spray cleaning (with a hot soda solution delivered by spray nozzles 2), spray rinsing and drying operations.

63. STRAIGHTENING

In some cases articles and tools distorted and warped in the heat-treating process can be corrected by a straightening operation. For this purpose the flat articles and tools are placed upon a completely even base plate and then straightened by means of copper hammers, used to make a series of light strokes in the desired places. Steel hammers are not used because they leave impressions on the surface of the steel. Long rod-shaped tools are also straightened with light hammer strokes, upon being laid on two supports.

The straightening of tools is an operation demanding great skill and by far not always producing satisfactory results. Breakage of a part of the tools to be straightened can hardly be avoided in this operation. This proves once more that it is necessary to take all possible measures to prevent the distortion and warpage of articles and tools in the heat-treating process, or at least to reduce them to the minimum.

The straightening of rolled stock and large articles of structural steels is a much easier task, because special straightening machines, such as roll straighteners and power or hydraulic straightening presses have been developed for this purpose.

Chapter XII

CONTROL OF HEAT-TREATING PROCESS

64. PRODUCTION CONTROL ORGANISATION

Heat-treatment is a complex process. In the majority of cases the heat-treated parts should conform to very high quality standards within specified limits, these standards being extremely exacting in some cases. Even slight deviations from the established manufacturing process, let alone faulty workmanship, result in articles and tools not conforming to specifications, and being rejected because of poor quality.

Articles and tools are sometimes found not to specifications because the wrong grade of steel was used to manufacture them, i. e., not that grade that was specified in the working drawing. Another cause of poor quality resulting in rejections may be the unsatisfactory structure of the original steel, though its chemical composition may correspond to the specified grade. Examples of such nature have been referred to many times in the previous chapters of this book.

It is therefore necessary to ensure that the original steel, being a raw material about to be converted into a finished product, meets the standard specifications.

The second cause of rejections lies in deviations from the established manufacturing process. Pertinent examples have been already referred to. For instance, an excessive hardening temperature results in overheating the steel, whereas a too low hardening temperature or a too high tempering temperature lead to decrease in its hardness. A wrong composition of the nitriding gas used in the nitriding process may result in a brittle nitrided case. To prevent these and similar deviations *an adequate control of the manufacturing process must be provided.*

Thus, it is necessary to provide the following kinds of control inspection:

- 1) incoming steel inspection;
- 2) manufacturing process control;
- 3) control inspection of finished product, i. e., heat-treated articles and tools.

Who is in charge of such control? First of all, of course, such control should be carried out by the personnel performing heat-treatment, i. e., by workmen, foremen and process engineers. They must thoroughly control the process, both by supervising the thermocouples, rheometers and clocks used in production, and by inspecting the articles immediately after they have been quenched and testing them

for hardness. Efficient work is not imaginable without self-control, but self-control cannot be always fully objective. Therefore, at each plant there is a special *inspection department* responsible for the control of raw materials, adopted manufacturing process, and quality of the finished product.

Control of quality of the finished product, carried out by inspectors who are employees of the inspection department, differs from the control as effected by workers of any other department of a plant in that it is considered to be final; the product accepted by the inspection department as good is passed to user departments, whereas that not accepted as being clearly bad is, in accordance with the degree of non-conformity with the specifications, either worked over into a useful form or rejected. Decisions as to whether the finished product is good or bad, arrived at by the inspection department, are final and cannot be annulled. Like the productive employees, inspectors are made responsible for the quality of the finished product.

65. INCOMING STEEL INSPECTION

The original steel used to manufacture articles and tools must be accepted on the results of special investigations that may be grouped as follows:

- 1) chemical composition of the steel;
- 2) its microstructure;
- 3) austenitic grain size;
- 4) hardenability.

All steels are inspected as to their chemical composition, whereas the microstructure is examined only for some grades of quality and high-quality steels, the original structure of which may considerably affect the properties of heat-treated articles. Such steels are those intended to be cold-stamped, including tool steels, ball bearing steels and some others.

The steel is tested for its chemical composition and crystalline or grain structure at the respective steel-making plants. The results of such a testing are indicated in certificates or documents specially prepared for each lot of steel certifying that it has met the technical requirements of a standard. When there is such a certificate and the steel is stored according to the technical requirements, there is no need for its further testing at the machine-building plants. To prevent intermixing in storage of steels of different grades, each grade must be stored separately, either in separate stacks or in different shelves. The grade of steel and certificate number are indicated in a special tag appended to one of the rods in the given lot or to one of the pipes or wire coils. In sheet stacks the grade

is marked on the uppermost sheet, either by means of marking irons or with indelible ink. In an extreme case the grade may be indicated simply with chalk, since the steel must not be left without any qualifying mark because memory alone cannot be relied upon.

To prevent steels of different grades from becoming mixed the so-called coloured marking is used, consisting of painting the ends of rods with certain colours, different for each grade of steel. But as even the best organisation of storekeeping cannot fully prevent the steels of different grades from becoming mixed, each should undergo *chemical analysis* to determine its composition before being used to manufacture heavy-duty articles and tools. In such checking chemical analysis there is no necessity to determine the content of all the elements and impurities that may be present in the steel; it is required only to determine the content of carbon and those alloying elements that are the chief factor in determining the characteristics of the metal and, consequently, the grade of the steel. Therefore such checking of the steel chemical composition may be carried out with spectroscope, or steelscope, or even with spark test. Every heat-treatment operator should learn to use the spark test for determining the grade of the steel.

Checking the original steels by microscopic examinations aims to determine:

- 1) actual grain size in structural steels;
- 2) kind of pearlite in tool steels, whether lamellar or globular;
- 3) presence and nature of the non-metallic inclusions;
- 4) presence of the stringy or striped structure, indicative of the non-uniform distribution of carbides;
- 5) depth of the decarburised layer.

In the existing State Standard specifications for rolled stock of carbon structural steels, there are no special requirements as to the steel structure. As for a quality structural sheet steel intended for deep drawing, the State Standard lays down special technical requirements of structure concerning the grain size, banded structure and the cementite occurring freely. These features of the steel structure are evaluated on the basis of the corresponding multi-point scales that are appended to the State Standard on the method for determining the microscopic structure of a quality sheet steel. For different grades of steel the allowable values are specified, depending upon the per cent reduction in deep drawing required.

The presence and nature of non-metallic inclusions, banded structure, lines and carbide non-uniformity are determined from longitudinal microsections, while the depth of the decarburised layer, from transverse microsections. Both longitudinal and transverse microsections may be used to determine other peculiar features of the steel structure, such as grain size, form of pearlite, etc. Non-etched

microsections are ordinarily used to determine the amount and nature of non-metallic inclusions. Methods for determining the depth of the decarburised layer have been already discussed in Chapter V.

Other features of the steel structure are evaluated either by describing it or by appending photomicrographs to the test certificates; or, which is the best and most objective, by indicating the point pertinent to the structure on the basis of the standard scales of microstructures available. Many plants work out their own scales of microstructures. There are several scales specified by the State Standards.

At the present time incoming steel is tested for *austenitic grain size* only for purposes of pure investigation. Not a single State Standard for steels specifies any technical requirements as to the austenitic grain size. This problem is reserved for the future. The method used for determining the austenitic grain size is described in Chapter II. Methods of carrying out the hardenability tests have been described in Chapter III.

66. CONTROL OF MANUFACTURING PROCESS

In any heat-treating process a most important factor is the temperature, and its control should therefore be given the greatest attention. Beside the *temperature*, the *time of heating* and the *holding time* at a given temperature must be controlled, and in gas case-hardening processes the *composition of the gas medium* should be controlled as well.

Temperature control. Temperature control is effected on the basis of readings of pyrometers, e. g.: 1) thermoelectric couples, or thermocouples, and 2) radiation pyrometers.

A thermoelectric couple, or thermocouple, is a commonly used device employed for measuring temperature differences. Thermocouples may be used to measure a wide range of temperatures, from room temperatures to very high ones. They may be employed for measuring temperatures both in furnaces and salt baths, bent-type thermocouples being the most adaptable for the latter. In heat-treating shops the most widely used are the following types of thermocouples: a) chromel-cupel*, for use at temperatures below 600°C; b) chromel-alumel, for use at temperatures up to 1000°C; and c) platinum-platinumrhodium, intended for high-temperature work. Platinum-platinumrhodium thermocouples, being more stable than the other types, are generally used for temperatures above 1000°C, the permanently sustained temperature not exceeding 1300°C, the maximum temperature for short-time use being 1600°C.

* "Cupel" is an alloy intended for use in thermocouples, containing about 40 per cent nickel, 1 per cent manganese, the balance being formed of copper.

When a thermocouple is used to measure the temperature, it is desirable that its hot junction should touch the surface of the heated articles, or, at least, be placed as near to them as possible. Otherwise the thermocouple will measure not the temperature of articles but the temperature of the furnace heating chamber. Particular care should be taken to prevent the hot junction of a thermocouple from being placed near to the heating elements in electric furnaces. If this is not done the thermocouple will give a considerably higher temperature than that of the articles heated, and if the heating process be conducted in accordance with readings from the thermocouple fitted in such a manner, the articles will be always found to be underheated.

Thermocouples are graduated at the cold-junction temperature of zero, while the temperature in the shop and especially in the immediate neighbourhood of heating furnaces is considerably higher than zero. Because of this the thermocouple readings will be less than the actual temperature, the difference being equal to temperature of the shop or, to be more accurate, to the cold-junction temperature. To eliminate such errors the electrodes of the thermocouple should be connected with a millivoltmeter by means of compensating wires. If this is not possible the thermocouple should be disconnected from the millivoltmeter, and the pointer of the latter must be adjusted from zero to "room" temperature by a corrector.

When measuring temperatures with a thermocouple, its readings have nevertheless to be checked by direct watching of the heat colours. Such a check is necessary, first, because the thermocouple, if incorrectly fitted, may give false results, as has been previously pointed out. Secondly, a thermocouple can never ascertain whether the heating is *uniform* or not, and uniformity of heating is no less important a factor than the temperature itself.

Heat colour	Temperature, °C	Heat colour	Temperature, °C
Black red	550 to 580	Orange, or free	
Blood red	580 to 650	scaling heat	830 to 900
Dark red	650 to 730	Light orange	900 to 1050
Dark cherry red	730 to 770	Yellow	1050 to 1150
Cherry red, or full red	770 to 800	Light yellow	1150 to 1250
Light cherry, or light red	800 to 830	White	1250 to 1300

Wherever a furnace is used for heat-treating processes requiring long periods of time, e. g., annealing, tempering and case-hardening—the time is more readily controlled not with clocks but by recording millivoltmeters. These recording millivoltmeters or simply recorders are capable not only of measuring continuously the temperature of a furnace at any given moment, but of recording it upon special charts (Fig. 134). For this purpose the recording instrument is

provided with some kind of marking device, consisting usually of pointer with pen and ink. From time to time a special device presses the pointer upon a chart graduated with reference to temperature and time, and the pen traces a dotted line upon this chart, being slowly moved by a clock device. As the chart moves very slowly and the pen is made to press upon it fairly frequently, a continuous line representing the change in temperature of a furnace with time is traced on the chart. The resulting diagram is a sort of certificate verifying that the heat-treating process has been carried out in accordance with the specifications.

Electric furnaces that must have a constant temperature maintained for a long period of time are provided with thermoregulators. A thermoregulator is a sort of millivoltmeter of a more complex design than the ordinary one. Apart from the pointer, the dial of a thermoregulator may be provided with two indicators, a maximum and minimum (Fig. 135). Suppose that a furnace must be adjusted to maintain the temperature within a range from 780 to 800°C. The maximum indicator should then be adjusted so that its reading is 800°C, and the minimum one to the reading of 780°C. As soon as the pointer reaches the reading of 800°C the electric circuit is closed, and a relay built into the thermoregulator is made to switch off the furnace. The furnace ceases to be heated and begins to cool down. As temperature of a furnace begins to decrease, the millivoltmeter pointer moves to the left on the dial. As soon as it reaches the minimum reading, another circuit is closed, and the second relay switches on the heating elements. The furnace starts to heat up, with the pointer moving to the right in the direction of the maximum indicator, etc. It should be borne in mind that the indicators must not be fixed very close to each other without any special necessity, particularly in case of furnaces with a low heat inertia, or else the furnace will be switched on and off many times in rapid succession, with the thermoregulator made almost continuously to "reverse" with the result that its work will be interfered with.

To measure high temperatures of a magnitude such as are encountered in high-temperature salt baths, etc., thermocouples are not suitable, since their protective tubes, being immersed in the fused salt, rapidly burn out and the thermocouples become ruined. Therefore in such cases radiation pyrometers are recommended, being much more adapted for continuous measuring of very high temperatures. Such instruments consist of a case with a hot junction of small thermocouple 1 (Fig. 136) located on its optical axis. This hot junction bears a small blackened disc of platinum foil 2 attached to it, which is intended to concentrate the radiant heat of the object upon the couple. By pointing the optical axis of the radiation pyrometer toward a heated object, the temperature of which is to be

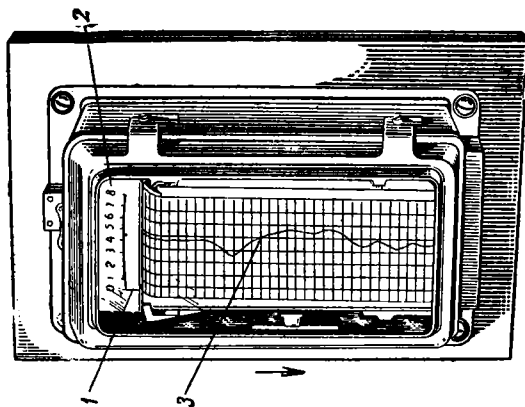


Fig. 134. Recording millivoltmeter with the temperature curve recorded on the chart paper.
1—arrow; 2—scale; 3—temperature curve on chart paper.

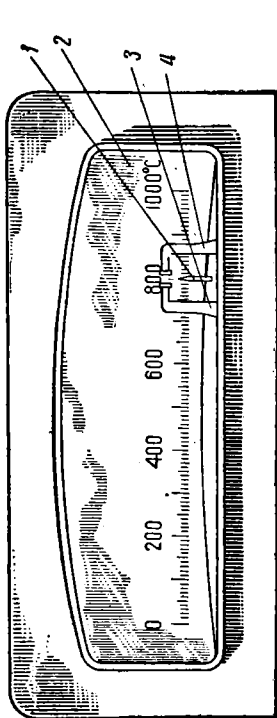


Fig. 135. Thermoregulator.

1—arrow; 2—scale; 3 and 4—indicators.

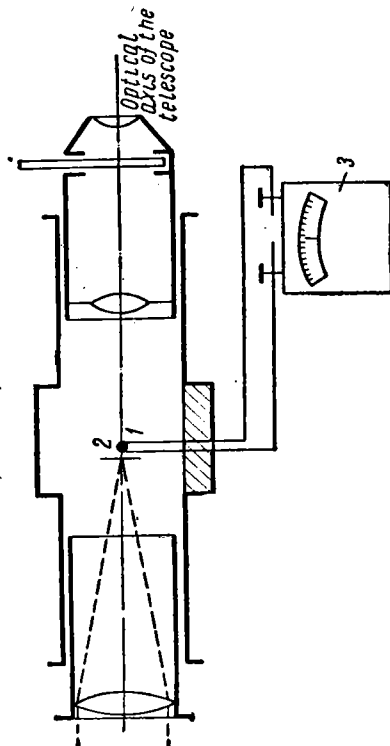


Fig. 136. Schematic diagram illustrating the design of a radiation pyrometer.

measured, and by focusing it accordingly (e. g., upon the surface of a salt bath), a portion of the radiant heat is made to fall upon the platinum disc, or "mirror", which concentrates the rays upon the hot junction of a small thermocouple, causing it to generate a current that can be measured by millivoltmeter 3. •

Radiation pyrometers may be used for automatic control of the temperature of the salt baths. This instrument, if properly fitted, adjusted and focused, can work automatically, i. e., without an operator.

A great disadvantage of radiation pyrometers is that their readings are dependent upon the degree of transparency of the air between the telescopes of the pyrometers and the surface of a salt bath being sighted. But the intense evolution of fumes from the surface of a salt bath at high temperatures produces screens of smoke or gases that absorb a portion of the radiant heat, thus causing the instrument readings to be lower than the actual values of temperature, sometimes by as much as 30 to 50°C. This fact should be borne in mind when measuring temperatures by means of radiation pyrometers. Remedies have been proposed, in particular methods for using compressed air in blowing off the smoke screen from the surface of a bath.

The so-called surface-type thermocouples are used to measure surface temperatures of articles heated below the red heat. In these instruments the hot junction is formed not with two wires but with a pair of fairly wide ribbons of dissimilar metals, connected together. This improves the heat conductivity from the article the temperature of which is to be measured to the thermocouple hot junction.

Lower temperatures, e. g., in quenching liquids, oil baths, etc., can be measured by liquid-in-glass thermometers (mercury-filled) for temperatures up to 350°C, or by electrical-resistance thermometers for temperatures up to 500°C. Low surface temperatures may be also measured (from 200 to 500°C) by means of the so-called thermocolours. These are substances of a complex chemical composition that change their colours suddenly when heated to some definite temperatures (usually different for various thermocolours). The part which is to be heated is painted in two or three stripes with a number of thermocolours that change colour at fairly close temperatures. As soon as the colour of a main thermocolour has changed in heating, the process of heating is stopped. Other thermocolours are used as controls in order to make comparisons.

Measurement of gas flow. In gas case-hardening processes the gas flows are measured by means of special devices, called rheometers (Fig. 137). Such meters consist of a conical tube, through which the gas to be metered is passed from below upwards. Inside the tube

there is a float. The higher the rate of gas flow, the higher the float will be raised. The instrument can be graduated directly in units of flow rate (e. g., in litres per minute), as the rate of flow per time unit (minute, hour) is equal to the product of the cross-section area (being a constant value for a given rheometer) and the velocity of the gas flow.

Measurement of degree of dissociation of ammonia. The degree of dissociation of ammonia applied in nitriding and gas cyaniding is measured with a special metering device called a dissociation meter (Fig. 138). This instrument consists of a glass vessel with a scale graduated in 100 divisions (the division value is irrelevant). The dissociation meter is

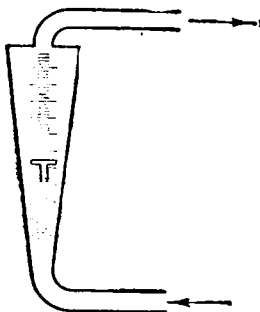


Fig. 137. Schematic diagram illustrating the design of a rheometer.

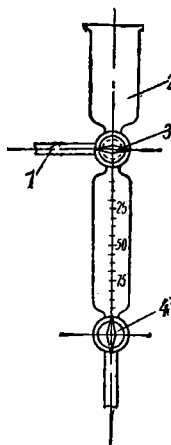


Fig. 138. Schematic diagram illustrating the design of a dissociation meter.

connected with the furnace by means of tube 1. To measure the degree of dissociation of ammonia both cocks (3 and 4) are opened and the exit gases from the furnace are passed through the dissociation meter. After some time has elapsed cock 4 is shut and cock 3 is turned to such a position as to prevent gas from passing through tube 1 and to admit water present in funnel 2 through a small opening. Water thus introduced into the dissociation meter absorbs all the ammonia that has been not decomposed, whereas hydrogen and nitrogen do not dissolve in it. The amount of water admitted into the meter is equal to the amount of ammonia being not decomposed; and thus the reading on the scale against the water level gives the degree of dissociation in per cent by volume.

67. QUALITY CONTROL OF HEAT-TREATED ARTICLES AND TOOLS (FINISHED PRODUCT)

Control of the final quality characteristics of heat-treated articles and tools is the main and the most important kind of control of quality of the finished product. The quality control applied to the fully heat-treated articles and tools aims to determine whether they conform to established specifications and standards. Therefore this kind of control is considered as the final and most responsible procedure in production control.

It is therefore clear that quality control of completely heat-treated articles and tools consists of determining whether the given items are within the specified limits of variability, i. e., whether or to what degree they conform to established specifications and standards. Specifications are developed on the basis of the following sources:

- 1) standards, for typical articles (such as forged shafts, axles, etc.) and tools (such as cutting tools, bits, screw taps, etc.);
- 2) plant technical requirements, e. g., for component parts and tools for which no State Standards have been developed;
- 3) drawings.

According to the technical requirements formulated in the Standards, specifications and drawings, the articles and tools are to be tested for:

- 1) hardness; testing for hardness is the most commonly used kind of inspection;
- 2) mechanical properties;
- 3) physical properties (e. g., magnetic properties for transformer steel, corrosion resistance for parts being nitrided to increase their resistance against corrosion, etc.);
- 4) depth of a case, for articles which have been case-hardened.

Independently of the technical requirements formulated in the Standards, specifications and drawings, the heat-treated articles and tools are to be *visually inspected* to see whether the items are free from such surface defects as cracks, distortion, warpage, scale, etc. To notice defects that either cannot be detected by a visual inspection (e. g., internal cracks) or are hardly detectable visually (such as fine surface cracks), special *crack-detecting methods of inspection* are used.

When various quality characteristics or properties have been determined by inspection, the main "*defective*" characteristics are established, on the basis of which the articles inspected are either accepted or rejected. Thus, for instance, in testing the mechanical properties of forgings the following properties are determined: tensile strength, yield point, elongation, reduction of area, impact

strength, and Brinell hardness. Of these six characteristics not more than three are usually accepted as the defective ones, e. g., yield point, elongation (or reduction of area) and impact strength. The values of these quality characteristics must be above those specified by the technical requirements, or else the forgings will be rejected. Other characteristics, qualified as non-defective ones, may be both above and under the values required by specifications. In effect, the non-defective characteristics need not be determined in inspection testing, but they are, as a rule, determined mainly for the sake of acquiring experience. However, as the time goes by, experience is gained and manufacturing processes perfected, a still greater number of these characteristics is being transferred into the category of defective. The Standards are now developing in the same direction.

There are a number of different kinds of inspection, the principal ones being *complete screening* and *sampling inspection* methods. In complete screening inspection each part manufactured is subject to inspection individually. Sampling inspection is performed on samples taken at random (usually 3 to 5) from lots of small parts. These samples are considered representative of the lot when it is not practical to inspect each piece. The amount of inspection to be done is sometimes expressed in terms of percentage, ordinarily from 2 to 10 per cent.

A number of similar articles manufactured from the same steel grade and subjected simultaneously to the same heat-treating process are considered as a *lot*. If the articles are hardened together, and tempering is done in two operations, it is considered that there are two lots of parts. The heavy-duty articles and tools are usually subjected to screening inspection, whereas the sampling inspection is performed on less important parts. Usually the sampling inspection is done on samples taken at random from lots of small parts, such as small-size springs, small carburised parts, etc.

If the samples are unsatisfactory, then it is permitted to take a *double* number of samples from the same lot and to subject them to a repeated inspection. If the test results are found to be satisfactory, the entire lot may be accepted. However, if the repeated inspection gives unsatisfactory results, it is possible to proceed in one of three ways:

- 1) The entire lot may be returned to the producing department to be corrected by an additional heat-treatment (e. g., if the hardness is too high, it can be decreased by repeated tempering).

- 2) A complete and reliable screening inspection of all parts may be carried out (if a portion of representative samples is found to be unsatisfactory in sampling inspection, while other samples prove to be good).

3) The entire lot may be rejected, when the additional heat-treatment is considered to be either impracticable or inadvisable (e. g., where there is a high degree of warpage).

The hardness of steel may be measured by several methods being now in common use. The principal methods extensively used are: Brinell's method (as performed on the TIII-type hardness tester); Rockwell's method (as performed on the TK-type hardness tester; and Vickers' method (as done on the TII-type pyramid hardness testing machine).

The *Brinell hardness tester* is employed for determining the hardness of comparatively soft articles which have been subjected to annealing, normalising or drawing. To determine the hardness of large billets and articles special flat specimens are cut off to be tested. If such specimens cannot be cut off, a portable device, called a "*Poldi*" tester (Fig. 139), is used. In this device the method of determining the hardness consists in delivering a hammer 1 blow simultaneously on the article 3 to be tested and on a standard specimen 2 of known hardness. Under such conditions the steel ball is indented on one side into a flat surface on the standard specimen and on the other side into the article to be tested. As already mentioned, the hardness of the standard specimen is known. So by measuring with a micrometer microscope the diameters of two impressions thus obtained (on the standard specimen and on article to be tested), the hardness of the articles may be computed by rule of proportion. If the hardness of the standard specimen is designated by H_s and the diameters of two impressions on the specimen and the article to be tested by d_s and d_a respectively, then the hardness of the article is calculated from the following formula:

$$H_a = H_s \left(\frac{d_s}{d_a} \right)^2. \quad (27)$$

On the basis of this proportion, there have been compiled special tables which are used in practice for reading directly the hardness numbers from the corresponding indentation diameters. In determining the hardness with the "*Poldi*" device errors amounting to 5 per cent of the value of actual hardness sometimes occur, but such a degree of accuracy is considered to be quite sufficient for large articles to be tested under shop conditions.

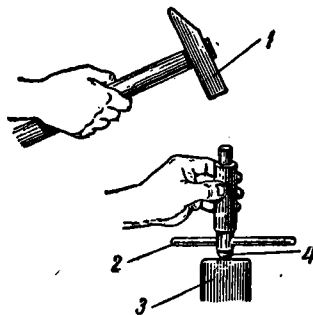


Fig. 139. Sketch illustrating the method of determining hardness on large articles by means of the portable hardness tester ("*Poldi*" tester). 1—hammer; 2—standard specimen; 3—part to be tested; 4—steel ball.

There are also *portable hardness testers* provided with standard scales reading directly in Brinell hardness numbers. Such testers consist of a rigid U-shaped frame (Fig. 140) which grips the article to be tested. In making the test a steel ball 5 mm in diameter is made to indent the smooth surface of a specimen by means of a screw-propelled base driven from a dynamometer. When pressure is applied to the predetermined load (usually about 750 kg) indicated on the dynamometer, and the diameter of the indentation is measured, the Brinell hardness number may be easily read directly from the usual tables.

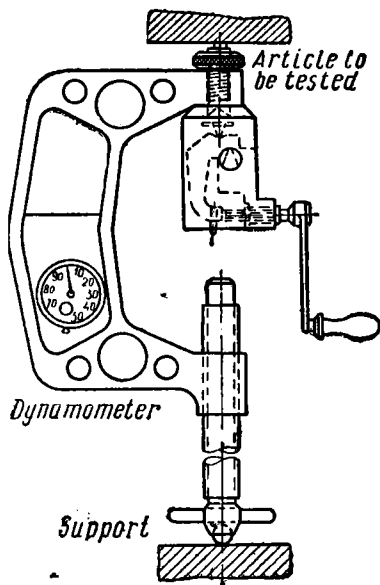


Fig. 140. Schematic diagram showing the design of a portable Brinell hardness tester.

The *Rockwell hardness testers* are used only to determine the hardness of the steel being hardened and tempered, thus having a fairly high hardness (not less than Rockwell C 25 to 30). It is not recommended to determine the hardness of annealed steels on the Rockwell tester, as the results obtained will be less correct than those produced by the Brinell hardness tester.

However, it is not always possible to set a tool on the base of a tester so that its hardness may be measured. For example, it is difficult to determine the hardness of bits, screw taps and springs.

In such cases it is necessary either to provide a special appliance designed to produce a reliable base for supporting the article while

testing by indenting a diamond penetrator (described as a sphere-conical-shaped diamond brale or cone, with a spherical apex of 0.2 mm radius forming an angle of 120°) or to ruin one to two tools from a lot by grinding two fields on their surface, one for making the test by indenting and the other serving as a supporting base; or, finally, to harden along with the lot of articles and tools a specimen of simple shape (flat or cylindrical) made from the same steel grade, using it as a sample for determining hardness. Such a procedure is followed, for instance, in determining the hardness of a lot of springs.

The *Vickers hardness tester*, employing a small diamond, cut and polished to form a square-based 136-degree pyramid, is extensively used to determine the hardness of very small and thin articles.

The loads applied in the test vary between 5 and 100 kilograms. Therefore all the indentations produced in the test are geometrically similar. (One of the advantages of this machine is that it is suitable for any degree of hardness.)

The so-called calibrated files with a defined hardness may be used for determining qualitatively the hardness of hardened tools. For example, if such a file is calibrated for a hardness of Rockwell C 60 to 62, it can easily scratch tools with a hardness less than Rockwell C 60 but cannot scratch those with a hardness above Rockwell C 62. In order to determine the hardness of tools by means of the scratch hardness tests, a number of files calibrated for different hardnesses is needed.

As for the control of *mechanical properties*, this question will be dealt with in detail in Chapter XV, where the heat-treatment of forgings for shafts will be considered, being selected as the most representative for this kind of inspection. The quality control of springs will be also discussed in Chapter XV, in the paragraph on the heat-treatment of springs.

The depth of a carburised case may be approximately (roughly) judged from fracture of test bars, or witnesses, placed in the carburising box with the pieces to be carburised (see Chapter X). If greater precision of measurement is desired than can be obtained by this method, transverse microsections may be examined by microscope, the case depth being measured with a scale in the eyepiece of the instrument or by some other optical method. A microscope is also used to determine the depths of nitrided, cyanided, calorised and other cases obtained in case-hardening processes.

PART THREE

HEAT-TREATING PROCESSES

Chapter XIII

PRODUCTION PROCESS ORGANISING

68. OPERATIONS INVOLVED IN THE HEAT-TREATING PROCESS

The heat-treating process applied to any article or tool is made up of a sequence of separate operations performed in a certain order, i. e., 1) preliminary operations in preparation for production; 2) main operations; 3) finishing operations; and 4) finished product control.

Preliminary preparation of articles and tools for heat-treatment is accomplished in different manners, depending upon the type of articles to be treated and the kind of heat-treating process to be performed. Preliminary operations include placing the objects on trays, packing the articles in boxes with iron cuttings and chips, in carburising pots, or in special baskets for gas carburising, nitriding or gas cyaniding in pit-type furnaces, etc. Parts to be heated in salt or saltpeter baths are first placed in special appliances or bound with wire. In selective carburising, as applied to articles of which only certain parts are to be carburised, the parts not to be carburised *are coated* with some substances that resist penetration by the carburising gases, e. g., they may be protected by copper plating, preferably electroplating. In the selective hardening process, as applied to articles and tools of which only certain parts are to be hardened, there should be also protection of those parts that are not to be hardened, such as threaded parts, auxiliary holes in dies, etc., or else quenching cracks may develop in them.

By *main operations* are understood the proper heat-treating operations, such as annealing, normalising, hardening, tempering, ageing, case-hardening, etc.

After the main operations have been completed, the heat-treated objects are cleaned from scale, degreased, and, if necessary, straightened. These *operations* are termed *finishing*.

At the end of the heat-treating process, the finished articles are subjected to *inspection*, being tested to determine their hardness, mechanical or any other properties as specified. In some cases the

heat-treated articles are inspected twice in succession, as the case may be in heat-treating tools, for instance, when the hardness is generally determined both after hardening and after tempering.

69. PRODUCTION PROCESS PLANNING

The planning phase of the heat-treating process starts with the study of drawings of the article to be treated. In the drawing the grade of steel used to manufacture the article is indicated, as well as the specifications that should be met by the article in as-treated condition. The technical requirements stated in the drawing usually specify the hardness desired; less frequently, mechanical or any other physico-chemical properties are also specified. For articles to be case-hardened not only the hardness is specified, but also the case depth, e. g., the depth of a carburised case. In some drawings it is specified that the article should be subjected either to local hardening or selective carburising.

The planning of the heat-treating process consists first of all in determining what manufacturing operations are to be applied to the given article. In most cases where the technical requirements as stated in the working drawing specify only the hardness no difficulties are experienced in choosing the heat-treating operations. For example, tools are to be hardened and tempered; springs, hardened and drawn-back; steel castings, usually annealed; sheet blanks for deep-drawing, normalised, etc. Some difficulties are experienced in selecting the kind of heat-treatment to be applied to articles for which certain mechanical properties are specified in the drawing; such articles may be either normalised or hardened and subsequently tempered. In such cases the problem is solved by comparing the mechanical properties, as specified by the technical requirements in the drawing, with the properties of the given steel grade in as normalised or refined by heat-treatment condition, known either from the literature sources or from data obtained in the plant practice. As a general rule it may be accepted that articles of structural carbon steels should be normalised, whereas those made from alloy steels should be hardened with subsequent tempering.

When the type of the final heat-treatment has been selected, it is necessary to determine the *sequence* of the machining and heat-treating operations, i. e., the place of the heat-treating process in the whole sequence of manufacturing operations deemed necessary for producing the given article.

Articles not to be machined, such as leaf springs, hot-wound springs, most of the parts produced by cold stamping, etc., are shipped directly for assembling after the heat-treatment is completed.

For articles to be machined the sequence of heat-treating and machining operations is determined chiefly on the basis of the hardness values obtained as a result of heat-treatment. If this hardness is comparatively low (not higher than 270 to 300 H_B) and does not interfere with the machining, the heat-treating process is then done before the machining operation. If, on the contrary, the heat-treating results in a high hardness (above 270 to 300 H_B), then the machining must be performed in two stages: preliminary machining to be done prior to the heat-treatment, and the finish-machining to be performed after the article has been fully heat-treated. In the preliminary machining, or roughing operation, small allowances are left on the article to be removed by the finish or final machining operation. The same procedure is followed in the case when the hardness is not too high but the article should have, after machining, a high surface finish and accurate dimensions, e. g., in the case of bolts, shafts, splines, etc. In such cases the operation sequence must be: 1) preliminary machining; 2) heat-treatment; 3) finish-machining; 4) assembling.

Though articles with a hardness of over 200 to 250 H_B can be machined, this operation presents some difficulties and, what is most important, reduces the cutting speed. Low feeds and cutting speeds are conducive to shortening the service life of cutting tools; in such cases, therefore, the heat-treatment is usually also done in two stages, the preliminary heat-treatment being performed to improve the machinability of the steel, while the final treatment is intended to give the article the desired mechanical properties. In such a case the sequence of manufacturing operations will be: 1) preliminary heat-treatment; 2) preliminary or rough machining; 3) final heat-treatment; 4) final machining; 5) assembly. Such an operation sequence is adopted in manufacturing crankshafts, rotors, dies and punches, heavy-duty gears, machine-tool spindles, etc.

At this (first) stage of production process planning the question is also being solved as to whether it is necessary to include in the manufacturing process such heat-treating operations as homogenising to remove the non-uniformity in chemical composition (as is done, for instance, on alloy steel castings), normalising to prepare the steel structure for hardening and to diminish distortion and warpage, etc.

When the general procedure to be followed in the production process has been established, methods of carrying out the main heat-treating operations are chosen. If, for example, hardening is selected as the main heat-treating operation, then it should be decided what type of hardening is the most appropriate for the given case, either ordinary hardening, interrupted hardening, isothermal hardening, or surface hardening with heating by gas flame, or by high-frequency electric current, etc.

The manner in which this problem can be solved depends on many factors, of which the principal are the following. First of all, the selection of treatment depends upon the availability of suitable machinery and equipment in the shop. Secondly, it depends on a lot size, i. e., on the number of similar articles to be treated. Some methods of heat-treatment require the devising of a special outfit. For instance, for surface hardening through heating by high-frequency electric current it is necessary to manufacture a special inductor designed to be used with the given article. If a lot size is small, then the designing and manufacture of a special equipment may prove to be too costly and not worth while from the economical point of view. If, on the contrary, a lot of articles to be treated is of a sufficiently large size, then in most cases the making of such a special outfit is to be considered as quite worth while, since it permits the introduction into the industry of manufacturing processes which are more technically refined and more easily applicable.

Selection of one of the possible methods must be corroborated by an adequate cost calculation. To this effect the cost of auxiliary materials, outfit and equipment, fuel or electric power, as well as of labour is determined for each method of operation. The method accepted as the best under the given conditions should facilitate the carrying out of the production process in the most economical manner in terms of materials and labour; moreover, it must enable the cutting down of the amount of manual work to be done in the process and the introduction of a greater degree of mechanisation and automation, and must be capable of completing all the work in the most direct manner and shortest time to meet a projected completion date.

The best conditions for introducing advanced and more refined manufacturing processes are available under mass and large-lot types of production. In such types of production any refinement of the production process, however slight it may appear, which aims to cut down the time required to complete or to decrease the quantity of work involved, contributes greatly to raising its operating efficiency. The economic efficiency involved in introducing advanced manufacturing processes in the job-type (individual) and small-lot type of production is less. However, even in these types of production the introduction of more refined manufacturing processes help to improve the quality of the finished product and to increase productivity. In any case, the heat-treating operator is expected to introduce more advanced manufacturing methods into any type of production wherever technically practicable and economically advisable.

Thus when the manufacturing operations of the production process have been selected, the process engineer begins to work on the third

phase of production planning, involving the elaboration of a complete list of operations, including preparation, finishing and inspection. In this phase of planning the process engineer also prepares a list of appliances necessary for carrying on separate operations and distributes assignments of outfits to be designed by the engineers.

The fourth phase of planning the production process consists of determining the factors pertaining to the temperature of heating, heating time and holding time of heating, methods of cooling, etc.

At this stage of planning the selection of process equipment required for the practical carrying out of the elaborated manufacturing process is also made. This mainly implies the selection of suitable heating furnaces. In choosing the furnaces the process engineer bases his decisions on the shape and size of articles to be heat-treated, the size of the lots and the degree of furnace charging required.

In working out the detailed production process he is guided mainly by his theoretical knowledge and practical experience. In each heat-treating shop a great deal of experience is gradually accumulated, and in elaborating the detailed sequence of procedures of the heat-treating process to be performed on a new article the results of heat-treating similar articles already obtained are taken into account. Such a method of working out the detailed manufacturing processes is one of the most reliable because it is based on practically verified data. This method, however, should not be overestimated: frequently in heat-treating shops a method is progressively developed and eventually adopted that may be a good but not the best, and a good tradition may thus turn into a routine. Therefore, the experience gathered in shop practice must be always used only after a critical examination of the pertaining data.

However, there are also cases reported where even the special handbooks cannot provide data needed to determine in advance the detailed sequence of operations for a heat-treating process, as, for example, the case may be when some alloy steel article of intricate shape has to be treated. In such cases test pieces are heat-treated first, and then thoroughly tested and inspected in the plant laboratory. The results of such inspection and testing are used to correct the adopted heat-treating process so as to meet the specified technical requirements.

70. OPERATION CHARTS AND INSTRUCTION LISTS

When the manufacturing process has been determined on the basis of the drawings, bills of materials and specifications, it is recorded in *operation charts*, prepared by a process engineer. These charts are designed to instruct heat-treaters as to the sequence of opera-

tions and necessary tools, outfits, and time allowances that should be conformed with in carrying on the heat-treating process.

In each plant such operation charts are prepared in different forms, depending upon the various local conditions. But whatever plant may prepare these operation charts their main contents must be always the same.

Table 21 shows an example of such an operation chart.

In the operation charts there must be indicated:

- 1) article name;
- 2) drawing No.;
- 3) sketches of article layout, with indication of main (external) dimensions, as well as of those dimensions that have a direct bearing on the heat-treating process, specifying what part should be carburised (if selective carburising has to be done), or what part must be hardened (if the hardening should be a local one);
- 4) grade of steel;
- 5) specifications of hardness desired, mechanical properties, depth of carburised case, etc.;
- 6) sequence of operations;
- 7) equipment to be used under the particular conditions of the heat-treating shop;
- 8) appliances and outfits to be used in the particular heat-treating operation; there is, of course, no necessity to state the standard devices that are employed in any heat-treating process, such as thermocouples, tongs, etc.;
- 9) detailed sequence of main procedures in the heat-treating process, complete with basic operation data, e. g., initial temperature of a furnace, when it is being charged with articles to be heated; either the total time of heating and holding at the given temperature (when heating in salt baths) or only the time of holding (when heating large articles in batch-type furnaces); cooling environment (or method of cooling), i. e., furnace, open air, water, oil; temperature of the cooling liquid (usually indicated in cases where it must differ from the room temperature);
- 10) number of articles to be taken at random for sampling inspection by testing for hardness or mechanical properties;
- 11) classes of work, time allowances either in all operations or only in main procedures (if the preliminary and finishing operations are paid for by lots rather than individually for each article treated), and wage rates on operations according to the class of work and standard times established.

Operation charts are signed by the process engineer who worked out the detailed sequence of operations in the production process and by the head of the engineering department, and are confirmed either by the chief process engineer or by the chief metallurgist.

Table 21

Operation Chart No.

Article: <i>ratchet wheel</i>		Grade of steel: 15-20		(Sketch of part)					
Drawing number:		Specifications: 1. <i>Carburise to depth of 1.0 to 1.2 mm</i> 2. <i>Harden for Rockwell C 60 to 62</i>							
Manufacturing process									
Item No.	Operations	Equipment	Appliance	Procedure to be followed			Class of work	Time allowance	Wage rate
				Temperature of furnace	Temperature of operation	Time of heating or holding			
1	<i>Packing in box</i>	—	<i>Carburising box</i> —	—	—	—	—	<i>40 pieces</i>	
2	<i>Carburising</i>	<i>Batch-type furnace No. 2</i>	—	<i>700° C</i>	<i>900° C</i>	<i>8 hours</i>	—	<i>Time to be corrected after inspection of test bars</i>	
3	<i>Control of carburising depth</i>	—	—	—	—	—	—	—	
4	<i>Dumping of box</i>	—	—	—	—	—	—	—	
5	<i>Cleaning of articles</i>	—	<i>Wire brush</i>	—	—	—	—	—	

Continued

Item No.	Operations	Equipment	Appliance	Procedure to be followed					Class of work	Time allowance	Wage rate
				Temperature of furnace	Temperature of operation	Time of heating or holding	Cooling	Note			
6	Hardening	Batch-type furnace No. 4	Box with iron swarf	760 to 780° C	760 to 780° C	To be determined by tempering colour	In water (20 to 30° C)	—			
7	Tempering	Oil bath	Net	150 to 170° C	150 to 170° C	One hour	—	—			
8	Degreasing	Rinsing tank	—	—	—	—	—	—			
9	Testing for hardness	Rockwell tester	—	—	—	—	—	2 pieces from lot			
10	Inspection of case depth	—	—	—	—	—	—	To be performed in metallographic laboratory on 1 sample from lot			
Prepared by process engineer		(Signature)	Head of engineering department	(Signature)	(Date)	Rate setter	(Signature)	(Date)	Approved by plant chief process engineer	(Signature)	(Date)

The second important type of technological document is the *instruction lists*. These differ from the operation charts in that they refer not to some particular article but to a group of identical articles, e. g., "Instruction list on heat-treatment of springs", or to some particular kind of heat-treatment, e. g., "Instruction list on nitriding operation"; or, finally, to running some particular equipment unit, e. g., "Instruction list on work with cyanide baths".

Unlike the operation charts, the instruction lists set out in greater detail all the steps of each particular operation. However, they are, of course, less concrete than the former, e. g., in an operation chart there is indicated the precise time of heating of a particular article (e. g., 15 minutes), while in an instruction list there is only stated the time allowance for each unit (e. g., 1.5 minutes for each millimetre of cross-section).

The operation charts are, as a rule, prepared only on articles manufactured in mass or lot types of production; for non-standardised articles produced on the job-type basis, or for the small-lot production, the operation charts are set up only on the heavy-duty items, e. g., forgings for shafts. Instruction lists must, however, be set up for technical purposes for all operations to be accomplished in a heat-treating shop.

Chapter XIV

HEAT-TREATMENT OF ROLLED STEEL

71. VARIETIES OF ROLLED STEEL

Rolled steel embraces rods of various shapes (rounds, squares, hexagons), bands, strips, angles, channels, double-tee bars, rails, plates, sheets, ribbons, tubes and wire. Rods, bands, angles, channels, bars, beams, and rails are known under the common term of section steel. Light section steel is a common term given to wire and ribbons and other semi-finished products (nets, bolts, nuts, nails, etc.).

All classes of the rolled steel are obtained by hot-rolling. Rolled steel which is rolled when hot and not given supplementary rolling or drawing in a cold condition is called the *hot-rolled* steel. Some hot-rolled rods, bands, strips, tubes and wire are additionally cold-drawn. This steel is termed cold-drawn steel (sometimes also bright-drawn or sized steel). *Cold-drawn* steel comprises small-section rods, strips and bands, thin-walled tubes and wire of small diameter. Some hot-rolled sheets and ribbons are additionally cold-rolled, being then called *cold-rolled* steel.

Some hot-rolled steel and all cold-drawn and cold-rolled steel are heat-treated. The heat-treatment applied to these two kinds of the rolled steel is performed in different manners.

72. HEAT-TREATMENT OF HOT-ROLLED STEEL

As a rule hot-rolled carbon structural section and sheet steels are not subjected to a special heat-treatment. The heat-treatment of rolled stock of carbon structural steels is performed simultaneously with the rolling operation itself.

Hot-rolled sheet steel intended for deep drawing must be heat-treated. Low-carbon sheet steel (up to the grade 20 inclusive) is either annealed or normalised so as to obtain the fine-grain structure, while medium-carbon sheet steel (grades 25 and up) should be annealed for the globular pearlite to form. Thin sheet, roofing sheet and the so-called pickled (annealed and subsequently pickled) steel must also be annealed to obtain the fine-grain structure.

Thin sheet steel for electrical purposes, e. g., dynamo and transformer sheet steel, is annealed at 850 to 860°C, with a long holding at this temperature followed by a slow cooling. The annealing of electrical steel is performed in order to improve its magnetic properties.

In annealing the processes of decarburising, partial graphitisation of the residual carbon and growth of the ferritic grains all take place. They are apt to decrease the coercitive force and, consequently, the losses due to hysteresis and eddy currents. This is apparently the only case where coarse-grain structure is intentionally produced in steel.

It is absolutely necessary to heat-treat *hot-rolled stock* (e. g., sectional steel, sheets, and tubes) made of tool and other high-carbon steels. The purpose of heat-treating such hot-rolled steel is to obtain the globular pearlite structure, conducive first of all to improvement of the machinability of the steel. Hot-rolled steel is therefore subjected mainly to annealing, while some highly alloyed or complexly alloyed steels—to high temperature tempering.

The annealing process applied to the hot-rolled steel is represented in the diagram of Fig. 141. For all steels of this class annealing

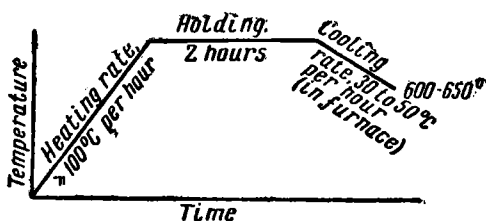


Fig. 141. Diagram illustrating the annealing of hot-rolled steel.

is performed in the same way, differing only in the heating temperature and the holding time of heating. Some highly alloyed steels (e. g., grades X9C2, X10C2M, EBK30, etc.) are annealed at about 900°C, which process is nevertheless to be termed a high temperature tempering since the A_1 point for these steels is located at temperatures above 900°C. Hypereutectoid, chromium alloyed steels, such as grades 11X15, 11Γ, XBF, are annealed at about 800°C.

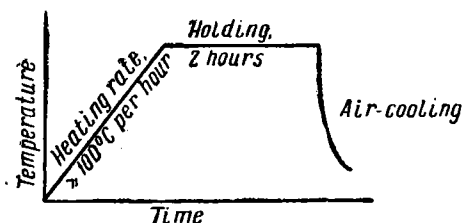


Fig. 142. Diagram illustrating the high-temperature tempering of hot-rolled steel.

temperature tempering. This applies first of all to steels with austenite having a high stability within the range of $A_{r'}$ -transformation, such as grades 18XHBA, 25XHBA, 5XHM, 5XFM. High-temperature tempering is preferred for steels in the structure of which there are carbides tending towards intensive coagulation or coalescence with a rise of temperature, such as the grades XB5, 3X2B8, 4XB2C. When these steels are annealed the carbides grow considerably in size and are difficult to dissolve in the solid solution when subsequent hardening of the billets made of such a rolled stock is carried out.

The diagram in Fig. 142 illustrates the process of high-temperature tempering performed on these steels: the tempering temperature ranges from 640°C for the grade 18XHBA to 850°C for grades X9C2, X10C2M.

73. HEAT-TREATMENT OF COLD-DRAWN AND COLD-ROLLED STEELS

As distinct from hot-rolled steel, which need not necessarily be heat-treated in some cases, as for example is the case with carbon steels, cold-drawn and cold-rolled steels must always be heat-treated regardless of the steel grade.

Hot-rolled steel is heat-treated only once, i. e., after rolling operations have been completed. On the other hand, cold-drawn and cold-rolled steel are heat-treated in three stages of the manufacturing process:

- 1) preliminary heat-treatment before cold-drawing or cold-rolling;
- 2) intermediate heat-treatment during the processes of drawing or rolling;
- 3) final heat-treatment when the drawing or rolling operations have been completed.

The *preliminary heat-treatment* consists of annealing to obtain globular pearlite or high-temperature tempering to form the sorbite structure. The preliminary heat-treatment applied to rolled stock of high-speed steels, which are to be drawn, has some peculiar features; it is conducted as follows: annealing at about 860°C, followed by slow cooling (at a rate of 10 to 20°C per hour) to 760°C; holding at that temperature for 5 hours, with a subsequent slow cooling (at a rate of 10 to 20°C per hour) to a temperature of 680°C. This annealing process is followed by the so-called carbide tempering, intended to dissolve part of the fine carbides, distributed at the grain boundaries, in the ferrite, and to retain them in the solid solution by rapidly cooling. High-speed steel subjected to carbide tempering is capable of withstanding higher degrees of reduction in cold working than steel which has not been so tempered.

The *intermediate heat-treatment of rolled stock*, performed at intervals between the separate passes, consists of recrystallisation annealing, the temperature of which ranges from 660 to 820°C according to the grade of steel treated (this applies to highly alloyed steels of type 1X18H9).

The *final heat-treatment* of cold-drawn section steel or cold-rolled sheet steel is selected in accordance with the technical requirements of the mechanical properties of a steel; it may be either the above-mentioned recrystallisation annealing, or full annealing, normalising, or hardening (austenitic steels only).

The essential peculiarity of the heat-treating of cold-rolled and cold-drawn steels which distinguishes it from the treatment of hot-rolled steel is that it is necessary to protect the steel thoroughly against oxidation and especially decarburisation. While an oxidised layer can be removed by pickling, as is usually done after each annealing operation, the decarburised layer cannot be removed. To protect the steel from decarburisation it is preferred to carry out high-temperature tempering rather than annealing, to accomplish the annealing or tempering at temperatures as low as possible, and to conduct the heating in tubes luted with fire-clay (for sectional steel), in welded-up boxes (for sheet steel), or in furnaces provided with a protective gas atmosphere.

74. HEAT-TREATMENT OF WIRE

Wire is obtained from a hot-rolled steel rod by a cold drawing operation. The heat-treatment applied to a cold-drawn wire does not essentially differ from that of a cold-drawn steel.

It is interesting to discuss the heat-treatment of spring wires of grades B, II, H, OBC and BC, made of the higher carbon (0.7 to 0.9 per cent) steels. The heat-treatment of a spring steel wire,

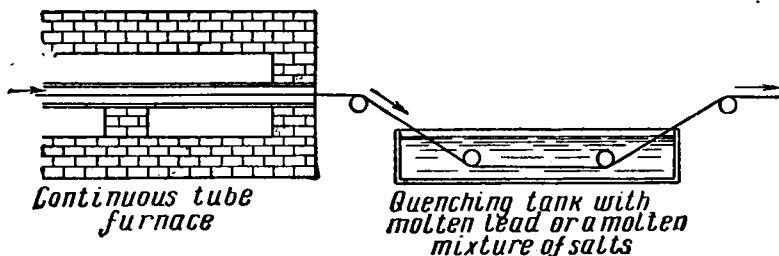


Fig. 143. Schematic diagram illustrating the wire patenting process.

the so-called patenting, is a process consisting in heating the material to a point considerably above the critical temperature and cooling through the critical temperature at a comparatively rapid rate. In practice there are several ways of administering the treatment. Thus the wire may be continuously passed through a long tube furnace (Fig. 143), kept at a temperature above the A_1 point by about 80 to 100°C, heated there and then cooled by drawing into a bath of a relatively cold lead or saltpeter maintained at a temperature of 450 to 550°C, i. e., the wire is subjected to the *isothermal hardening*. As a result of patenting, the structure of the wire is transformed into sorbite representing a step in the transformation from austenite to pearlite, in which the grains are very small and the iron carbide is distributed in very finely divided form in the ferrite. The patented wire is then subjected to cold drawing. In manufacturing wire of small diameters the patenting process is performed several times in succession.

Patented and subsequently cold-drawn steel wire combines a very high tensile strength with high plasticity, which gives it the ability to withstand cold winding, as in the manufacture of springs.

It should be noted that a patented wire has higher tensile strength and elasticity than a wire that has been hardened and tempered after cold drawing. In the first case the toughening obtained in the heat-treatment is supplemented by the strain hardening in the wire-draw-

ing process, whereas in the second case this strain hardening is considerably decreased by the recrystallisation which takes place in heating for hardening. Therefore it is not advisable to harden springs wound from patented steel wire.

Chapter XV

HEAT-TREATMENT OF MACHINE PARTS

75. HEAT-TREATMENT OF STEEL CASTINGS

Steel castings are made of carbon or of alloy steels. In the machine-building industry most steel castings are made from low- and medium-carbon steels of grades 15JI, 20JI, 30JI, and 35JI. In grade designation numbers for steels, the two-digit figures indicate the average carbon content in "points", or hundredths of 1 per cent, while the letter "JI" denotes that these steels belong to the class of cast steels, i. e., steels suitable for making castings. In chemical composition the carbon steels used for castings differ in general only slightly from hot-rolled quality carbon steels.

Steel castings must in most cases be subjected to a heat-treatment. Only low-quality castings made of low-carbon steels (grades 15JI and 20JI) may in some cases be used without being heat-treated. However, this practice cannot be considered a correct one and may be permitted in some instances only as an exception.

Steel castings must be heat-treated for the following reasons:

- 1) the original structure of steel castings is nearly always *coarse grained* (Fig. 144), i. e., structure resulting from an intensive overheating. This coarse-grain structure is characterised by poor mechanical properties. In such a steel the impact strength is especially low, as it is clearly shown in Table 22 where the mechanical properties of a steel of grade 20JI before and after annealing are compared;
- 2) steel castings that have been not heat-treated *have a poor machinability*, resulting from high hardness. At first glance this may seem strange; if the castings have been slowly cooled in sand mould, why should they be hard? This is explained by the fact that a coarse grain austenite is more liable to supercooling than a fine-grain one; therefore when the coarse-grain austenite decomposes, a finely dispersed ferrite-cementite conglomerate is formed;
- 3) in steel castings, even if they are slowly cooled in sand moulds, *high internal stresses are always developed during the solidification and subsequent cooling of the metal*. This is because comparatively few castings have a uniform section thickness; the steel castings usually consist of heavy sections in conjunction with thin ones.

Therefore the separate sections of castings do not solidify simultaneously, nor do they cool uniformly. And where the temperature is distributed unevenly, internal stresses are unavoidably set up.



Fig. 144. Photomicrograph showing the structure of cast steel, 100X.

Table 22
Mechanical Properties of 20J Grade Steel
Before and After Annealing

Mechanical properties	Numerical values	
	Before annealing	After annealing
Tensile strength, kg/mm ²	38.3	43.7
Elongation, %	15.9	25.4
Reduction of area, %	20.3	50.2
Impact strength, kgm	1.3	6.0

It is clear, therefore, that the most practicable method of heat-treating steel castings is full annealing. In fact this treatment may solve simultaneously all three problems, i. e., it may refine the grain structure, remove the cooling stresses set up in casting and thus improve the mechanical properties, particularly ductility, and make the steel easier to machine.

Steel castings should be annealed only after all the sink-heads, gates, risers, and fins attached to the casting have been cut off. This is usually done with an oxy-acetylene flame. As the zones immediately adjoining the planes of cutting are subjected to an intensive heating they are liable inevitably to be overheated. Hence if the sink-heads, gates, fins and risers were cut off after annealing, the steel structure of some portion of the casting body would be detrimentally affected.

The steel castings are charged for annealing either in a cold furnace or at least in a furnace cooled down to a temperature not exceeding 300 to 400°C. After the temperature of a whole charge of castings has become uniform (this takes about 1 to 2 hours, depending upon the size of a charge), they are slowly heated at about 100°C per hour to the annealing temperature, which is selected in accordance with the steel grade used for the castings. At the annealing temperature the steel castings are held for a period of time determined by the following calculation: 1.5 to 2 minutes for each millimetre of the maximum section thickness of the casting.

After annealing the castings are slowly cooled, usually in the furnace, to a temperature not exceeding 300°C; further cooling may be conducted in the open air.

With low-carbon steel castings (grades 15JI, 20JI) it is preferred to carry out *normalising*, sometimes also called air quenching, rather than annealing. The normalising operation has also to be applied to higher carbon steel castings when their mechanical properties, as developed by annealing, prove to be lower than those specified by the technical requirements. After annealing such castings should be reheated to a lower temperature and cooled slowly, an operation called high-temperature tempering or drawing. This treatment, usually conducted at a temperature of 600 to 650°C, is intended to relieve internal stresses set up by cooling and to improve the machinability of the steel.

Annealing alone is sometimes found to be insufficient for heavy castings of carbon steels. In such castings there is obtained a very coarse dendritic structure, with the carbon distributed in the austenitic grains extremely non-uniformly, so that an ordinary annealing cannot equalise the carbon concentration. Therefore in order to remove the intracrystalline segregation *homogenising annealing* is resorted to, this being an operation performed at 900 to 1000°C for a period of several hours, usually from 5 to 10. Although the temperature of the homogenising annealing is too low to equalise the carbon concentration in the austenitic grains completely, it has to be applied, since heating to higher temperatures may cause the castings to become distorted to a considerable extent because of the excessive plasticity and softness of the steel at such temperatures. After homogenising the steel castings are subjected to a secondary, *ordinary annealing*.

Castings made of alloy steels of the pearlitic class are homogenised (in the case of heavy castings), normalised, hardened and highly tempered. Castings of alloy steels of the austenitic class (e. g., high-manganese steel castings) should be hardened.

76. HEAT-TREATMENT OF FORGINGS OF SHAFTS AND AXLES

According to the cross-sections and mechanical properties desired, forgings of shafts and axles are made of quality carbon steels of grades 30, 35, 40 or of various types of structural alloy steels. Table 23 shows the standard mechanical properties for forgings of shafts and all-forged rotors of steam turbines as specified by the former Ministry of Heavy Machine-Building.

In accordance with the mechanical properties desired, forgings of shafts are classified into six main groups. The higher the mechan-

Table 23

**Minimum Values for Mechanical Properties of Steels Used for Forgings
of Shafts and Axles and for All-Forged Rotors of Steam Turbines**

Group of forgings	Tensile strength, kg/mm ²	Yield point, kg/mm ²	Elongation, %	Reduction of area, %	Impact strength, kgm	Recommended steel	
						Group of steel	Grade
I	52	28	19	40	4	Carbon steel Nickel steel	35, 40, 35H
II	58	35	17	40	5	Nickel steel Chrome-molybdenum steels	40H, 35XM, 34XM1A
III	65	50	15	40	6	Chrome-molybdenum steels	34XM, 34XM1A
IV	72	60	15	40	6	Chrome-nickel-molybdenum steels	34XH1M, 34XH2M
V	82	70	14	40	6	Chrome-nickel-molybdenum steels	34XH1M, 34XH2M, 34XH3M, 34X2H3M
VI	87	75	13	40	6	Chrome-nickel-molybdenum-vanadium steel	34XH3MΦ

Note: With the exception of grades 35 and 40, all steels recommended are non-standardised.

ical properties specified, the more highly alloyed steels should be used to make the forging. This needs no further explanations. It should be noted also that Table 23 indicates what grades of steels should be selected for particular groups of forgings.

The final selection of the steel grade is up to the plant making these forgings. The main consideration is the desired mechanical properties, whereas the chemical composition of the steel is optional.

The size of forgings has also its bearing on the selection of a steel grade; the larger the maximum size of a forging, the more highly alloyed steel must be selected. This is explained by the fact that as the content of alloying elements increases, the hardenability of the steel also increases, and consequently the mechanical properties desired will be secured in the greater portion of a forging.

The process of manufacturing shafts consists of the following operations:

Forgings of I to II Groups

forging
annealing
rough turning
normalising
tempering
testing
finish machining

Forgings of III to VI Groups

forging
isothermal annealing
rough turning
hardening
tempering
testing
finish machining

Sometimes forgings of shafts of group I, made from carbon steel, are not rough-turned, all machining operations being performed only after testing. Therefore there is no need for an annealing operation. But alloy and particularly complexly alloyed steels must necessarily be annealed after forging, as the annealing prevents the development of minute internal flaws, known as flakes.

Flakes are usually very numerous, thin, twisted cracks that are detected only on the machined surfaces of finished articles or on etched transverse specimens cut off from the special allowances. These flakes may be of various lengths, ranging from fractions of a millimetre up to several dozens of millimetres. The width of these haircracks is measured in hundredths of a millimetre. A steel with such minute internal flaws as flakes has very low ductile properties. Hydrogen is said to cause flakes in steel, although this is not yet definitely established. However, hydrogen does dissolve in the liquid steel during melting and remains in solution even when the steel has already solidified. With a decrease in temperature the solubility of hydrogen in steel also decreases, since it is thrown out of the iron space lattice and accumulated in microscopic pores. When its pressure has become very high, being measured in hundreds and thousands of atmospheres, the hydrogen is said to cause bursts, or ruptures, in the metal, thus forming the above-mentioned flakes.

It has been found that a slow cooling of steel (at a rate of 10 to 20°C per hour) after a forging operation, especially in the temperature region around 200°C and below, completely prevents flakes from developing in forged articles. Due to such a delayed cooling the steel is said to become resistant to the formation of flakes, independent of how many times it may be subsequently heated and cooled. Apparently, in slow cooling the hydrogen has time to diffuse toward the surface of the steel and escape into the atmosphere.

Instead of slowly cooling, forged steel may be isothermally annealed at temperatures of 640 to 660°C. The holding time at that temperature depends on the steel grade and size of forging:

Diameter of Forging, mm	Steels of Grades 34XH1M and 34XH2M	Steels of Grades 34XH3M and 34X2H3M
	24 hours	60 hours
up to 300	36 "	90 "
300 to 500	60 "	120 "
500 to 700	96 "	180 "
700 to 1,000		

The diagram of Fig. 145 illustrates the heat-treatment of forgings of shafts. The forgings are placed in a furnace heated to a temperature not exceeding 500 to 650°C (usually 300 to 400°C). Heating to that temperature is conducted at a rate depending on the steel grade and the maximum diameter of the forging treated (see Chapter V). At a temperature of 600 to 650°C the forgings are held for a period of several hours (2 to 5 hours) for the temperature to become equalised in the whole cross-section. The further heating of forgings to a

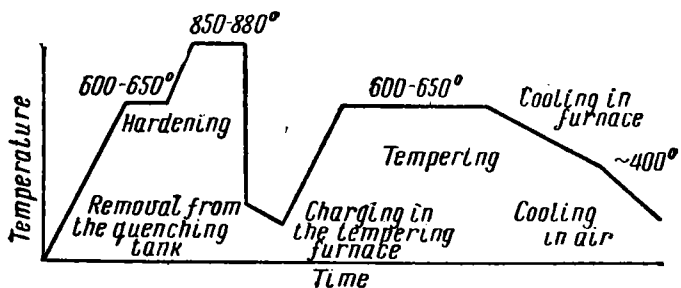


Fig. 145. Diagram illustrating the heat-treating process for forgings of shafts.

normalising temperature or hardening temperature may be conducted at the highest rate of which the furnace is capable, as was pointed out previously (see Chapter V). At the present time some plants are using the accelerated heating technique consisting of placing carbon and low-alloy steel forgings in a furnace heated up to 950 to 1000°C. Such accelerated heating may be applied to good effect to forgings that have been already previously annealed and are thus freed from internal stresses.

The normalising or hardening temperatures are determined according to the grade of steel treated (see Table 15). With complexly alloyed steels of non-standardised grades, the hardening temperatures are as follows:

Grade of Steel to Be Treated	Hardening Temperature, °C
34XM, 34XM1A, 34XH1M	850 to 870
34XH2M, 34XH3M 34XH3MΦ	840 to 860

The holding time of heating at the normalising temperature is determined by formula (12):

$$z_{\min} = (0.5-1.0) H_{\min}$$

(in the given case $H = D_{\max}$).

With low-alloy steels of a low hardenability there is no necessity for a long holding at the normalising or hardening temperatures for the temperature to become equalised in the whole cross-section of a forging, for its core will not be hardened anyway.

After normalising it is sufficient to cool forgings to 400 to 450° C, upon which they are placed in the annealing furnace.

In hardening forgings of complexly alloyed steels they should be deeply cooled, so that their core is cooled below 300° C. In such a case, the A_r -transformation will be prevented in the central portion of a forging, and the martensitic or A_r'' -transformation will take place, with the result that the core of the forging will develop high mechanical properties after having been tempered. But if the core of a forging is not cooled below 300° C at the end of its cooling in a quenching tank, then the A_r' -transformation may occur, with the result that the ductile properties of the steel, and particularly its impact strength, will be drastically decreased.

To lower the temperature of the central portion of a forging below 300° C it must be delayed in a quenching tank, with the time of holding forgings 400 mm in diameter in the oil bath extending up to 40 minutes, and forgings 600 mm in diameter up to 160 to 200 minutes.

Forgings to be tempered should be placed in a furnace heated to a low temperature, say 150 to 300° C, varying inversely with diameter of the forging and the degree of alloy content of the steel. Heating forgings up to a tempering temperature is conducted at a slow rate ranging from 20 to 80° C per hour, the lower value being applied to large diameter (800 to 1,000 mm) forgings of complexly alloyed steels, and the higher values to smaller diameter forgings of carbon and low-alloy steels.

The tempering temperature is to be determined either from Table 24 or from the experience gathered in the shop practice.

The holding time at a tempering temperature is usually 2 to 3 times longer than that at a hardening temperature, being computed on the basis of 2 minutes for each millimetre of the largest diameter of the shaft treated. After tempering the forgings are slowly cooled at a rate of 20 to 50° C per hour to a temperature of 100 to 300° C, the length of the cooling process increasing with the size of the forgings treated. The further cooling may be done in the open air.

The heat-treating process applied to large forgings must be further verified as to the times of heating, holding, and cooling. There is much evidence that the speeds of heating and cooling applied in shop practice are selected too slow in many cases, whereas the holding times are generally too long. Examination and correction may be accomplished only on the basis of thorough and elaborate experiments. However, such experiments with large forgings are both very complicated and extremely costly.

Table 24

Tempering Temperatures for Forgings

Grade of steel	Tempering temperatures in ° C, suitable to obtain yield point values as given below						
	30 to 35	35 to 40	40 to 45	50 to 55	60 to 65	75 to 80	80 to 85
40, 40H	630 to 660	620 to 650	600 to 630	—	—	—	—
40X	—	630 to 660	610 to 640	580 to 620	—	—	—
34XM	—	—	640 to 660	590 to 640	580 to 620	—	—
34XH1M	—	—	—	610 to 650	590 to 640	570 to 590	—
34XH3M	—	—	—	630 to 650	600 to 640	570 to 600	550 to 570
34XH4M	—	—	—	—	—	580 to 600	550 to 580

Forgings of shafts are subjected to the following kinds of testing: 1) testing for mechanical properties; 2) testing for flakes, and 3) testing for the residual stresses.

Forgings are classified into five groups according to their quality classes. Accordingly, the State Standard specifies a number of standard *mechanical tests*, as indicated in Table 25.

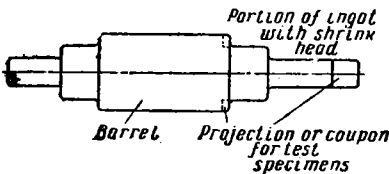


Fig. 146. Drawing of shaft forging.

Forgings of shafts are sometimes included in group IV, but usually they are classed in group V. All large forgings of complexly alloyed steels are referred only to group V.

Specimens for mechanical testing are cut out from the special allowances or "samples" (Fig. 146). For comparatively small class I forgings of shafts (up to 500 mm in diameter) of carbon steels it is sufficient to test one sample, usually from the end of the forging which corresponds to the ingot end where the shrink head was located. With more heavily stressed and larger forgings specimens are cut from both ends. As for very heavily stressed and very large forgings,

Standard Tests for Forgings

Table 25

Group of forgings	Main characteristics of group		Values of mechanical properties necessary for acceptance of lots
	Kind of control testing	Conditions for making up a lot of forgings	
I	Without testing	Forgings of the same steel grade	—
II	Testing for hardness of samples for each lot	Forgings of the same grade of steel having been similarly heat-treated	Brinell hardness (H_B)
III	Testing of each forging for hardness	Forgings of the same grade of steel having been simultaneously heat-treated	
IV	Testing of each forging for hardness and of representative samples for other mechanical properties	Forgings of a steel of the same heat having been simultaneously heat-treated	Yield point or tensile strength (left to the choice of user), elongation, reduction of area, and impact strength
V	Testing of each forging for mechanical properties	Each forging has to be accepted individually	

such as shafts for turbines and turbo-generators, etc., the specimens are usually cut from both ends and from the barrels.

The specimens for testing are cut from projections known as coupons, left at both ends of a forging, at a distance of $1/3$ of radius from the forging surface. From each coupon two specimens are prepared for a rupture test and two specimens for impact tests. As a rule the specimens are cut out in a longitudinal direction, i. e., in such a manner that their axes are parallel to the axis of the forging. For

shafts that are subjected in service to high tangential stresses (or torsional stresses), forgings are tested upon the so-called *transverse test specimens* (Fig. 147), the long axis of which is tangential to

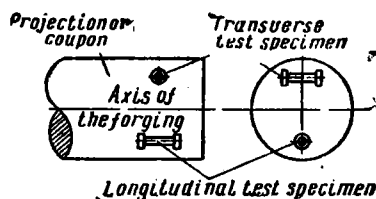


Fig. 147. Sketch illustrating the location of longitudinal and transverse specimens in the forging of a shaft.

the cross-section of the forging. The transverse or tangential test specimens are cut out from a projection or coupon left on the barrel of the shaft near the end of the ingot at which the shrinking head was located. As the transverse specimens are situated perpendicularly to the direction of the fibres of the forging, the values of mechanical properties obtained from the tangential specimens are lower than those obtained from the longitudinal specimens, the former values being the lower, of course, the higher the per cent reduction of the steel while forging.

The forgings of heavy-duty shafts, made of complexly alloyed steels which are the most sensitive to such internal flaws as flakes or haircracks, are *tested for flakes*. This testing consists of pickling with special reagents the neck journals and ends of shafts, as well as the end faces of barrels in the case of rotors. The testing for flakes consists of the following operations: 1) grinding, cleaning, and degreasing of the surface to be tested; 2) etching for a period of 10 minutes in a 15 per cent (by volume) ammonium persulphate water solution; 3) subsequent etching for a period of 5 to 10 minutes in a 10 per cent (by volume) solution of nitric acid; 4) two inspections of the etched surface, the first being done 10 minutes after etching, and the second not earlier than a day after it.

For especially heavy-duty forgings the *internal stresses are determined*, this testing is performed as follows: A portion of the barrel ("test") 20 to 25 mm wide is thoroughly machined and ground. Then the diameter is measured in four directions. Measurements are taken on a machine-tool after the shaft has been completely cooled down. The part of the barrel where the measurements have been taken is then cut out in a form of a ring, this operation being performed at minimum cutting speed and minimum feed, and with abundant cooling. Next second measurements are taken of the same diameters that have been previously measured. If there were internal stresses present in the forging, the diameter values measured before and after the cutting out of the ring will not be identical.

The internal tangential stresses are determined by a simplified formula:

$$\sigma = E \frac{\Delta D}{D}, \quad (28)$$

where E —modulus of elasticity;

ΔD —average value (of four measurements) of a change in diameter D .

Let us assume that $D=600$ mm and that $\Delta D=0.12$ mm.

Then

$$\sigma = 20,000 \times \frac{0.12}{600} = 4 \text{ kg/sq mm.}$$

For forgings of groups I, II, and III the internal stresses should not exceed 10 per cent of the real minimum value of the yield strength, while for forgings of groups IV to VI, they should not exceed 8 per cent of that value.

77. HEAT-TREATMENT OF SPRINGS

Helical twist springs may be made of carbon steels of grades 65, 70, 75, and 85, silicon steels of grades 50C2, 55C2, 60C2, 60C2A, manganese steels of grade 65Г, silicon-manganese steels of grades 55ГC, 55ГC, 60CГA, and some other more complexly alloyed steels, such as 60C2BA, 60C2H2A, 60C2XΦA, 60C2XA, 50XΦA, 50XГΦA. The most typical is silicon steel of grade 60C2, which is used for coiling most springs. It is comparatively cheap, since it does not contain expensive alloying elements, is easier to heat-treat than other spring steels, and has good elastic properties (e. g., high yield point). The heat-treatment of springs usually consists of quenching in oil, followed by tempering. In some cases, when the original steel has a coarse-grain structure, the springs are preliminarily normalised.

In heating springs for hardening special care should be taken to prevent them from changing their dimensions and shape. To achieve this the springs should be first of all heated in a horizontal position, for a spring being placed vertically for heating may deflect under the action of its own weight while hot. To prevent springs from becoming distorted the bottom of the furnace on which the springs are placed to be heated must be smooth and even. To prevent the small gauge wire, long springs from warping in heating they are fitted on mandrels made of sections of thin-walled pipe. To prevent the tightly wound springs (without clearances between the turns) from deflection in heating, they are firmly bound with a steel wire along the cylinder generatrix.

Springs should be hardened at the following temperatures:

Grade of Steel	Hardening Temperature, °C
65	840
70	830
75 to 85	820
65Г	830
50C2 and 55C2	870
60C2 and 60C2A	870
55ГC	820
55CГ	880
60 CГ and 60CГA	860
65C2BA and 60C2H2A	850
60C2XΦA	850
60C2XA	870
50XΦA and 50XГΦA	850

In heating springs for hardening it should be borne in mind that all steels alloyed with silicon are very liable to decarburisation. Therefore the steel must be held at the hardening temperature for as short a period of time as possible, calculated on the basis of 1 minute for each 1 mm of wire diameter. Moreover, measures should be taken to prevent decarburisation, though they be the simplest ones, such as covering the bottom of the furnace with charcoal or used carburising compound.

Manganese alloyed steels are very sensitive to overheating. In heating springs of manganese steel it is therefore necessary to control the temperature of the furnace very closely. As a rule springs should be quenched in oil, independent of whether the steel used to manufacture them is a carbon or alloy one. The springs must be immersed in the quenching tank vertically, with their long axis perpendicular to the surface of the quenching liquid.

Springs must be tempered in saltpeter baths. Tempering in batch-type furnaces results in the mechanical properties, included hardness, being non-uniform over the length of the springs. In some instances this is responsible for the breaking of the springs in testing, or, what is worse, when in service. Very serious attention must be paid to this fact.

Tempering temperatures are determined as follows, depending upon the steel grade:

Grade of Steel	Tempering Temperature, °C
60, 70, 75, and 85	480
65Γ	480
50C2 and 55C2	460
60C2 and 60C2A	460
55ΓC	480
55CΓ	460
60CΓ and 60CΓA	460
60C2XΦA	410
60C2XA, 65C2BA, and 60C2H2A	420
50XΦA and 50XΓΦA	520

The tempering time is calculated on the basis of $1\frac{1}{2}$ minutes for each millimetre of the wire diameter, being not less than $\frac{1}{2}$ hour in any case.

For springs wound from a wire 10 mm in diameter or less, a cold-drawn patented wire is usually used.

The patented wire has very high tensile properties; for example, grade B wire 1 mm in diameter possesses a tensile strength of 250 kg/sq mm, whereas the non-patented wire of the same grade 6 mm in diameter, 140 kg/sq mm. Yield point values are 0.8 to 0.9 that of the tensile strength values. In addition to the high tensile strength and

elastic properties of the patented wire, it possesses good ductility and toughness.

A great advantage of patented wire is that springs wound from it do not require hardening. Moreover, if a spring wound of patented wire were hardened its elastic properties might even become lower, as the heating for hardening is apt to remove cold hardening.

Springs wound of the patented wire are subjected only to tempering (preferably in oil baths or in furnaces with forced circulation of air at temperatures of 200 to 250°C) to relieve the internal stresses developed in the winding operation. In some instances, when

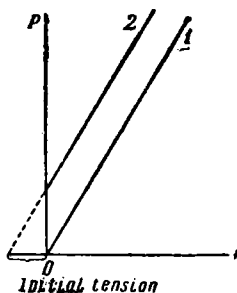


Fig. 148. Characteristics of a normal spring 1 and of a spring with initial tension 2.

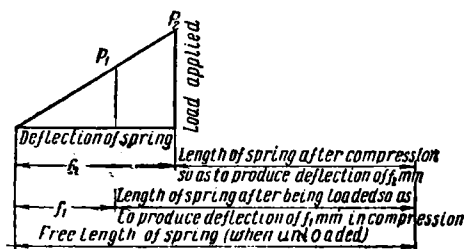


Fig. 149. Characteristics of a spring, with indication of "points" or positions where the spring is being tested.

extension springs are wound without any clearance space between each turn (with pitch or lead equal to the wire diameter), the turns are so tightly drawn together that some force P_i has to be applied to pull them apart. This force P_i is described as "initial tension" (Fig. 148), which causes the turns to be drawn tightly together. With the initial tension present in the spring, there is no direct proportionality between the load P applied and the extension f . To remove this initial tension such springs are recommended to be tempered at a higher temperature (400 to 450°C).

As a result of hardening and tempering springs should develop hardness ranging from 40 to 45 Rockwell C hardness numbers. But it is impossible to determine the hardness exactly by testing the spring itself. Therefore, the procedure adopted is as follows: to one of the springs of a lot to be treated a small, straightened piece (called a test piece) of the same wire the springs are made of is appended. This test piece is subjected to the same heat-treatment as the springs; it may therefore be used to test the hardness of the whole spring. Patented wire is usually not tested for hardness.

Heat-treated springs are subjected to tension or compression tests. If the spring is sufficiently long, these consist of its being extended or compressed in a testing machine to the value of deflection indicated in the working drawing, the load supported by the spring, which is read directly from the dynamometer scale of the machine (Fig. 149), being proportional to the deflection value. If the load is within the allowable limits, as specified by the drawing, the spring is accepted as good. Sometimes the drawings specify the load that

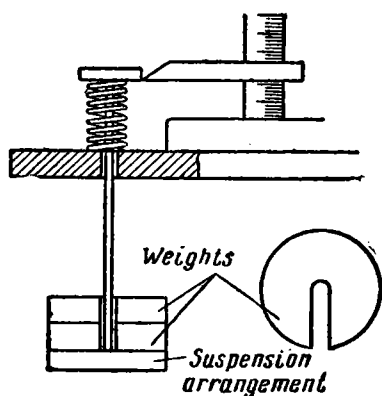


Fig. 150. Sketch illustrating the testing of small springs by means of loading with weights

should be applied to a spring for its extension or compression, with a deflection value to be determined in testing. If the deflection value is found to be within the allowable limits, the spring is again considered to be a good one. Heavy-duty springs are subjected to a complete screening inspection with each spring being tested individually; whereas springs to be used under less exacting conditions are subjected to a sampling inspection, where representative samples taken at random from lots of springs are tested.

Small-size springs are not suitable for a testing machine, so they are tested by means of load and height gauges in appliances as shown in Fig. 150. Springs which do not correspond to the deflection values specified in the drawings, when supporting the loads in testing (or vice versa), are rejected as being clearly bad. It should be noted, however, that this need by no means be the fault of the heat-treatment. Indeed the relation of the deflection f to the applied load P is given by the formula:

$$f = \frac{8nPD^3}{Gd^4}, \quad (29)$$

where n —number of effective turns, i.e., turns likely to be deflected;
 D —mean diameter of spring, i. e., outside diameter minus wire diameter;

G —torsional modulus of elasticity;

d —diameter of spring wire.

In this formula (29) there is only one value characteristic of metal properties, i. e., torsional modulus of elasticity G .

This value does not depend on the heat-treatment, as the modulus of elasticity (both Young's modulus and torsional modulus) is com-

pletely the same for steel in both annealed and hardened conditions. Thus the characteristic of a spring depends (for a given material) exclusively on its heat-treating conditions. Involuntarily the question arises as to why, in such a case, the springs are hardened? This is very simply explained: formula (29), like all the formulas for strength of materials, is correct only for elastic deformations and cannot be applied to plastic deformations. By hardening the spring the yield point of a steel is raised, i. e., the region of elastic deformations is widened, thus extending the ranges for application of formula (29). The diagram in Fig. 151 illustrates this very well. The straight line $0-1$ is characteristic of a spring in any condition, both as-hardened and as-annealed. The angle of inclination φ is a constant value for the given spring, as is evident from the formula:

$$\tan \varphi = \frac{P}{f} = \frac{d^4}{8nD^3} G \quad (30)$$

(f is substituted by its value, as obtained from formula [29]).

For an annealed spring formula (29) is correct up to the value of P_{ya} , corresponding to the yield point of an annealed steel. After the P_{ya} -value has been reached the annealed spring is characterised by a curve designated as "annealing". For a hardened spring formula (29) remains correct up to the P_{yh} -value, corresponding to the yield point of a hardened steel. After this value is attained the curve designated as "hardening" will be characteristic of the hardened spring.

Hence, if the testing proves that a test load supported by the spring does not correspond to the specified deflection value, the cause of this should be sought in deviations of the spring dimensions (n , D , d) from those given in the working drawing; either the spring has been wound from a wire of a wrong diameter, or an error in the number of effective turns has been committed in the winding.

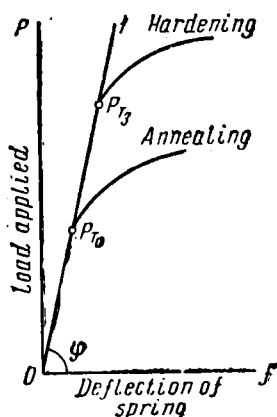


Fig. 151. Characteristics of annealed and hardened springs.

78. HEAT-TREATMENT OF GEARS

Gear teeth usually work under most severe conditions of service. They should possess very high strength, since they are intended for transmitting large torques, combined with the very high wear resistance necessary to prevent them from wearing away in service, for any distortion of teeth forms may interfere with the smooth action

of mating gears. So the steels that gears are made of must meet certain requirements.

Gears of relatively small power-transmitting capacity, i. e., those that work under low specific pressures and at low velocities, are made of carbon steels of grades 45 or 50. For heavier service conditions (higher unit stresses and high velocities at the pitch diameter) gears may be manufactured of alloy steels of grades 40X, 35XM, 50Г2, 40XH, etc. Finally, gears working under severe conditions and also likely to be subjected to shocks in service should be manufactured from carburised steels, especially alloy steels of grades 20X, 20X3, 12XH3A, 18XГТ, 18XHBA, etc.

Gears made from a plain carbon steel of grade 45 are hardened by quenching in water from a temperature of between 820 to 840° C, with subsequent high tempering (drawing) at a temperature of 520 to 550° C to obtain hardness of 220 to 250 H_B .

Gears made from the 40X grade chromium steel are hardened by quenching in oil from a temperature of between 830 and 840° C. The subsequent tempering may be conducted in different manners, as follows. When gears are intended for working under low unit stresses, they may be tempered at 600 to 650° C in order to obtain the Brinell hardness ranging from 230 to 260 H_B . But if the gear teeth are to be subjected to considerable unit stresses in service, the gears should be tempered at 180 to 200° C so as to obtain the Rockwell C hardness of 45 to 50 and prevent crushing of the teeth surface.

To increase the surface hardness of gears made from plain carbon and low-alloy steels capable of being refined, they may be subjected (as is actually done in many plants) to surface hardening treatment. Prior to the surface hardening, such gears are normalised or refined (quenched and tempered). The surface hardening treatment should be applied either to the whole surface layer of the gear circumference or, at least, to those face flanks of teeth that are most subject to wear in service.

The surface hardness of gears made from steels capable of being refined, i. e., responding to quench-and-temper treatment, may be also increased by cyaniding (by either gas or liquid) to a depth of 0.2 to 0.3 mm. The cyaniding operation is combined with hardening.

Gears from carburising steels are carburised to a depth of 1 to 1.5 mm. Gears, made of low-alloy steels (grades 20X, 18ГТ), are hardened immediately after carburising (from the carburising temperature), though some precooling is used thereby. Such a heat-treating procedure is both simpler and causes less distortion. But to gears made from more highly alloyed steels, such as grade 12XH3A, etc., this procedure cannot be applied, because the steel structure retains a very high proportion of residual austenite (up to 60 to 70 per cent) which is very difficult to decompose, and in sub-zero treatment cracks

often develop. Therefore gears of such more highly alloyed steels are cooled after carburising, with subsequent heating for hardening up to a temperature of 780 to 800° C, and quenching in oil. In some instances a sub-zero treatment of the steel at a temperature not lower than -60 to -80° C is performed to decompose the residual austenite, followed by tempering at 180 to 200° C to obtain the Rockwell C hardness ranging from 60 to 63.

If a carbide network forms during carburising, this must be followed by normalising before the gears are hardened.

Chapter XVI

HEAT-TREATMENT OF TOOLS

79. HEAT-TREATMENT OF CARBON AND LOW-ALLOY STEEL CUTTING TOOLS

Steel for cutting tools should meet the following main technical requirements. The hardness must be 60 to 65 H_C in as-hardened condition, which is considered quite sufficient for a cutting tool to engage easily in the material to be machined. Apart from the high hardness, these steels should possess a *high wear-resistance*, which is needed to prevent the cutting edges of a tool from becoming blunt by wearing away in service as a result of friction between the article and the chip.

Tool steels of high carbon content, being principally classed as hypereutectoid, conform to these requirements. Of the *carbon tool steels*, grades Y10 and Y10A, Y11 and Y11A, Y12 and Y12A are considered to be typical for cutting tools. Steels of grades Y13 and Y13A are hardly ever used for cutting tools, as they are found to be excessively brittle because of their high carbon content.

However, it is known that the carbon tool steels have considerable drawbacks that limit their uses. Since the main drawback of a carbon tool steel is its low heat-resistance, its hardness materially decreases as the cutting edge of a tool is heated up to 200 to 250° C, with the result that the cutting tool becomes blunt. Therefore at the present time carbon tool steels are used to manufacture only such cutting tools as are not subjected to high heating in service.

Whenever there is a great evolution of heat in cutting and a tool is consequently required to possess a high heat resistance, the *high-speed steels* should be employed. Still higher heat-resisting properties are possessed by the so-called *hard cutting alloys*. During recent times there has been increased use of still more heat-resisting materials, the so-called *ceramic* cutting materials. High-speed steels, hard

alloys, and ceramic materials are the main materials used for tools employed for high-speed machining of metals.

Some cutting tools are manufactured from the *low-alloy, high-carbon tool steels*. Though the heat resistance of such steels is somewhat higher than that of carbon tool steels (see Fig. 47), it is nevertheless considerably lower than that of heat-resistant high-speed steels. Therefore they are not used for cutting tools employed in high-speed machining of metals. The advantages of the low-alloy tool steels are of quite a different nature.

Large cutting tools cannot be manufactured of carbon tool steels because of their low hardenability and high critical quenching rate. Even when large tools are quenched in water the actual cooling rate is lower than the critical value, with the result that the hardness desired is not obtained. At best such hardening results in the formation of martensite in only a thin surface layer of a tool, and this may be crushed by the high stresses developed in cutting. But even when large tools of low-alloy tool steels are quenched in oil, the resulting cooling rate is higher than the critical quenching rate. The tools are hardened throughout and obtain a high hardness. In addition, quenching in oil almost completely prevents the formation of quenching cracks, which is a very considerable advantage in large cutting tools of intricate shape.

Low-alloy tool steels are also used for long and slim tools, such as long bits of small diameter, long and slender reamers, long screw taps of small diameter and long broaches, which should distort as little as possible in hardening. This can be achieved through quenching in oil.

Finally, a number of low-alloy tool steels, such as grades XΓ and XBΓ, are classed in a group of *low-distorting* steels.

During hardening steel of grade XB5 develops a very high hardness, usually in the range of 65 to 69R_C. This steel is called also "*diamond*" steel because of its ability to develop such a high hardness. It is used for cutting tools employed to machine articles of very high hardness, e. g., chilled cast-iron rolls. It should be noted that in spite of a very high hardness, the XB5 grade steel has a comparatively low heat resistance; therefore machining with tools of this steel should be conducted at medium cutting speeds.

The heat-treating processes applied to various tools of carbon and low-alloy tool steels are essentially similar one to another. Some differences are caused by the fact that certain tools (e. g., drills, milling cutters) should be hardened throughout, whereas others (e. g., screw taps, screw dies) are best not hardened throughout.

Let us consider first the heat-treatment of drills and then that of screw taps, since both are representative of the two groups of cutting tools.

Heat-Treatment of Drills. Drills should be hardened throughout in order to obtain a high hardness in the rib along the whole flute length, which is necessary for the drill to retain its high hardness after re-grinding operations. Therefore heats of high hardenability are particularly selected for carbon steel drills (points II or III, in the scale of Fig. 42). If steel of such a heat is unavailable, the drills must be first normalised for the lamellar pearlite to form, since this is conducive to higher hardenability, as was stated previously.

The heat-treatment process applied to drills consists of the following operations:

- 1) heating in a batch-type furnace up to a temperature of 550 to 600° C, followed by a final heating in a salt bath up to a hardening temperature corresponding to the steel grade (see Table 15). Small-size carbon steel drills may be directly heated in the salt bath without a preliminary heating in a batch-type furnace; the drills (especially those of the smaller sizes) should be heated vertically, being placed in the batch-type furnace with their shanks downwards, inserted in holes specially provided in fire-bricks;

- 2) drills of larger sizes (over 12 mm in diameter) made from carbon tool steels are transferred into oil after being quenched in water, whereas small-size drills (under 12 mm in diameter) made of carbon steels, as well as all drills of alloy steels, are quenched in oil. The drills are immersed in the quenching liquid strictly vertically; they should be removed from the oil at a temperature of 150 to 200° C (the oil fumes on the drill surface);

- 3) tempering is conducted in an oil bath either at a temperature of 150 to 180° C (for drills of carbon steels) or at 160 to 190° C (for drills of alloy steels) for a period of 1 hour;

- 4) degreasing;

- 5) straightening.

Heat-Treatment of Screw Taps. The heat-treatment of screw taps differs from that of drills in that it is desirable not to harden them throughout. This is preferred first of all because a non-through type hardening results in less distortion, with the thread pitch of a screw tap resulting more accurate. In addition the core of a screw tap is required to be tougher in order better to withstand the high torsional stresses that are set up in taps in service.

To provide the non-through hardenability of screw taps, the described heat-treating process is modified as follows. For the carbon steel screw taps heats of minimum hardenability are selected (corresponding to points I to II, as in the scale of Fig. 42). For alloy steel screw taps there is no possibility of different treatment in order to decrease their hardenability. First they must be quenched from a temperature corresponding to the lower limit of the range of hardening temperatures for a given steel, because the lower the hardening

temperature, the less will be the hardenability, as has been previously pointed out. Secondly, the screw taps must be heated in a salt bath without any preliminary heating in a batch-type furnace. Holding time in the salt bath should be minimum, and the screw tap must be quenched immediately after its colour becomes the same as that of the melted salt.

Screw taps, as well as screw dies, are tempered at a higher temperature (220 to 240° C) than drills. The selection of higher tempering temperatures for screw taps is aimed at increasing the toughness of the tool core.

Table 26 shows steel grades, as recommended by Standards and experience for various types of tools, as well as the standardised hardness values.

To conclude this paragraph, let us consider methods of heat-treating shanks of tools, the range of hardness for which should be Rockwell C 30 to 45. Such a hardness ensures high strength combined with a fairly high toughness. In order to give a tool different hardnesses, a higher one in the working part and a lower one in the shank, the following procedure is adopted. First, the shank is hardened and tempered, the latter at a temperature of 400 to 500° C; then the working part of the tool is hardened and tempered. Tools of carbon and alloy steels, having a low heat resistance (grades X05, B1), may omit the tempering of shanks, since in fact the shank will be tempered when the working part is heated for hardening. Another procedure may be also adopted, consisting of first hardening the tool throughout and then double-tempering it, low tempering being applied to the whole body of the tool, and local high tempering (at a temperature of 400 to 500° C) to only the shank. The local tempering of the shank can be effected through its short-time heating in either a salt or lead bath. Such is a procedure adopted in tempering shanks of threading dies in particular: the die is held in tongs and its shank brought into contact with the surface of a salt or lead bath. As soon as the shank has developed the blue temper colour, the tool is quenched in water to prevent the heat from being conducted further and avoid excessive tempering of the working, or cutting, part.

Instead of an additional tempering, the shank of a threading die is in some instances tightly wrapped with an asbestos cord; in tempering this part of the die will be cooled at a comparatively slow rate and thus will not be hardened.

Table 26

**Grades of Steels to Be Used for Main Types of Cutting Tools,
and the Standardised Hardness Values**

Tool name	Grade of steel	Dimension of working part, mm	Rockwell C hardness
Cutting tools	P18, P9	all dimensions	62 to 65
Twist drills	Y10A, (Y11A), Y12A, 9XC, (X, XBF), P9, P18	{ up to 10 over 10 up to 5 over 5	59 to 63 61 to 64 60 to 64 62 to 65
Counterbores	9XC, (X) P9, (P18)	all dimensions " "	61 to 63 62 to 65
Reamers	Y10A, Y12A 9XC, (X, XBF) P9, (P18)	{ 3 to 8 over 8 3 to 6 over 6	59 to 63 60 to 64 61 to 63 62 to 65
Screw taps	Y10A, (Y11A), Y11A Φ(9XC, XBF) P18, P9	{ all dimensions up to 6 over 6	59 to 62 61 to 63 62 to 65
Threading dies	9XC are allowed also Y10, Y12, Φ, XΓ (P9, P18)	all dimensions	58 to 62
Milling cutters: cylindrical, shell end, side milling, grooving, shank, angular, half-side milling, shank roughing	Carbon and alloy tool steels of grades (Y10A, Y12A, 9XC, XBF, X, XΓ, XB5)	all dimensions	61 to 64
	High-speed steels of grades (P18, P9)	all dimensions	62 to 65
Involute gear cutters	Y12A, 9XC	all dimensions	61 to 64
Milling cutters: cutting-off (or circular or disc saws), and keyway (or grooving cutters)	Y12A, 9XC	{ width up to 1 " over 1	58 to 62 60 to 64
	P9	{ " up to 1 " over 1	60 to 63 61 to 64

Note: Steel grades in brackets are indicated as supplementary to the grades as specified by the standards. They may be used whenever the main grades are unavailable.

80. HEAT-TREATMENT OF HIGH-SPEED CUTTING TOOLS

The heat-treatment of tools of high-speed steels differs in many respects from that of tools of carbon and low-alloy steels.

Because of the very low thermal conductivity of high-speed steel, tools made of it should be heated by stages. Small tools are first preheated in one furnace (preferably in a salt bath) to a temperature of 800° C, and then transferred to a high temperature furnace which is maintained at the proper temperature for hardening. Large tools of intricate shape must be heated in three operations carried out in

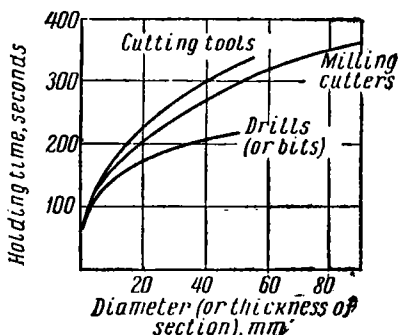


Fig. 152. Diagram to be used for determining the holding time of heating high-speed steel tools in a salt bath.

three separate furnaces, i. e., preheated to a temperature between 350 to 400° C in a batch-type furnace; preheated a second time up to about 800° C in a bath of molten salt; and finally, heated up to the proper hardening temperature. The holding time of heating at 800° C is calculated on the basis of 15 to 20 seconds for each 1 mm of the diameter or section thickness of the article to be treated when heating in a salt bath, or 30 seconds for each millimetre when heating in a batch-type furnace.

The final heating is best carried out in a bath of molten barium

chloride. In heating tools care should be taken to prevent decarburisation, which may be very considerable because of the high heating temperatures, and therefore the salt baths should be thoroughly deoxidised. Some tools, especially tools such as milling cutters with teeth that cannot be ground after hardening, are protected with a borax coating.

The holding time in a salt bath maintained at the proper hardening temperature may be determined from a diagram, developed by Y. A. Geller and represented in Fig. 152. To check the heating process the tool under treatment should be lifted from time to time above the bath surface for a period of 1 to 2 seconds. The heating may be considered as finished when the tool changes its colour to that of the molten salt, i. e., when it becomes only dimly visible on the bath surface.

Tools of high-speed steels are quenched either in oil, or in an air stream, or in a saltpeter bath heated to a temperature between 400 to 600° C. If the hardening is to be effected in oil, as is the most com-

mon case, the tool must first be precooled to a temperature between 900 to 1000° C (corresponding to the yellow temper colour) and only then quenched in oil. In hardening without precooling the tool, especially if it has a fairly intricate shape (e. g., milling cutters), may develop quenching cracks. The tool should be cooled in oil to a temperature of between 150 to 200° C (at this temperature the oil is slightly fuming on the tool surface), after which it is removed from the quenching tank and allowed to cool in the open air.

Small tools may be hardened by cooling in the air stream produced by fans.

Intricately shaped tools, such as milling cutters with sharp corners, junctions of heavy and thin sections, should be preferably quenched not in oil, but either in a saltpeter bath or in a bath of molten alkalis heated to a temperature of between 200 to 300° C, i. e., they are hardened by means of an interrupted quenching operation. The tools are held in such a bath for periods either of 30 to 40 minutes (at 200 to 250° C) or of 40 to 60 minutes (at 260 to 300° C). The subsequent cooling is conducted in the open air.

In large tools with junctions of heavy and thin sections quenching cracks sometimes develop even in hardening with interrupted quenching. In such a case, it may be advisable to quench the article in hot oil maintained at a temperature between 160 to 180° C, and to cool it further in the oil bath at a rate of 20 to 40° C per hour.

The range of hardness for properly hardened tools of high-speed steel is Rockwell C 60 to 63. If the tool is found to be harder than the 63 R_C after hardening, this fact is nearly always indicative of the tool being underheated in hardening, i. e., either the hardening temperature was too low or the holding time at the hardening temperature was not long enough. In such a case, the tool has to be tempered and its hardness tested again. If the tool has been underheated its hardness after tempering will be considerably decreased; such a tool has to be subjected to the repeated hardening.

This can be easily explained on the basis of the transformations, considered earlier, which take place in high-speed steels when they are being heated. In a case of underheating (Fig. 153), a maximum possible dissolution of carbides in the solid solution will not be obtained. If underheating of articles has been caused by too low a hardening temperature, the degree of dissolution of carbides obtained at that temperature will be lower than that which may be reached at a higher temperature; if, on the other hand, underheating had been caused by an insufficiently long holding at the hardening temperature, the carbides will not have been fully dissolved in the solid solution. In both cases the austenite will be less alloyed and, consequently, hardening results in the less retention of the residual austenite in the steel structure, with a corresponding greater amount

of martensite. This fact explains the somewhat increased hardness of a tool which has been underheated in hardening. The martensite obtained from the less highly alloyed austenite will certainly also be less alloyed and its resistance against tempering thus decreased. This is the underlying cause of the decrease in hardness of a tool which has been underheated in tempering.

If hardening of a tool results in a hardness below Rockwell C 60, this nearly always means that the tool has been overheated. Overheating produces a more highly alloyed austenite, the stability of which increases through dissolving a greater amount of carbides

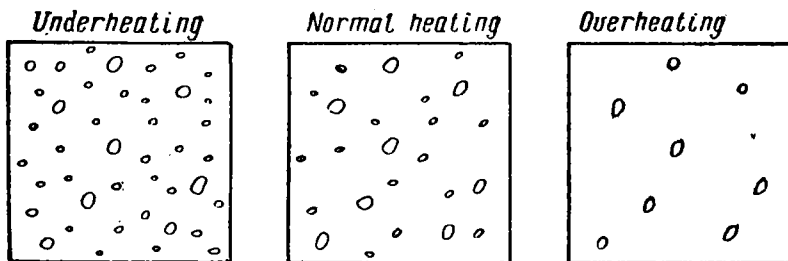


Fig. 153. Sketch illustrating the change in the amount of carbides in the structure of high-speed steel, depending upon the hardening temperature.

(Fig. 153); hence, in the structure of an overheated tool larger quantity of the residual austenite will be retained. Therefore the overheated tool has a decreased hardness. When overheating has not been very drastic, tempering will result in an increase in the tool hardness, the latter being capable of reaching its desired values.

The tool that has been underheated in hardening should be hardened again. In this connection it should be borne in mind that the tool must be annealed prior to the repeated hardening, or it will become very brittle after the secondary hardening because of an intensive coarsening of grains. A high-speed steel tool which has been hardened for a second time without an intermediate annealing shows very coarse bright grain grains in its fracture. Such a fracture has been termed a *coarse grain fracture*: separate coarse grains are clearly visible on its surface.

Tools of high-speed steels are tempered at a temperature of 550 to 560° C, usually in a series of operations, the first tempering being followed by a second, and sometimes even by a third and fourth ones. The third and fourth tempering operations are considered necessary whenever the tool has been slightly overheated in hardening, and single or double tempering operations are insufficient to obtain the desired hardness. Usually tools of the P18 grade steel are tempered

in two operations taking 1 hour each, and those of the P9 grade steel in two to three operations also of 1 hour each.

The range of hardness of tools of high-speed steels in as-tempered condition should be Rockwell C 62 to 65. .

81. CASE-HARDENING PROCESSES INCREASING THE LIFE OF TOOLS OF HIGH-SPEED STEELS

The service life of tools of high-speed steels, however long it may be, is nevertheless insufficient for the present high technical requirements of high-speed methods of machining that must be met. Therefore tool-makers, metallographers, and heat-treaters are trying to develop new methods of increasing the service life of cutting tools. At the present time a fairly great number of such methods are available, e. g., finishing, pickling, electrolytic polishing, chromium plating, electrospark machining, cyaniding, sub-zero treatment, sulphidising. Of all these methods, only the three last, falling into the group of case-hardening processes, will be discussed here.

The essentials of the *cyaniding process* have been considered in detail in Chapter X of this book. To supplement what has been said above, it remains to describe the peculiar features of cyaniding cutting tools.

Cyaniding can be applied only to the cutting tools manufactured of high-speed steels. Cutting tools made from carbon and low-alloy tool steels cannot be cyanided, since the cyaniding temperature is above the tempering temperature for such tools, and their hardness would therefore be decreased in cyaniding. Tools can, indeed, be heated for hardening in cyanide baths, thus combining the operations of high-temperature cyaniding and hardening; but this method is not widely used, mainly because, on the one hand, fairly deep cyaniding causes the cutting edges of a tool to become brittle and, on the other hand, shallow cyaniding is of little use because the cyanided case will be inevitably removed by the subsequent grinding of the tool. Low-temperature cyaniding is applied to tools of high-speed steels after they have been fully heat-treated and finish-machined.

There are three methods of cyaniding cutting tools: 1) liquid cyaniding; 2) gas cyaniding, and 3) solid or dry cyaniding.

Liquid cyaniding may be accomplished in cyanide baths of varying compositions, the most suitable for the purpose being the following (in per cent by weight):

First composition:	Per cent by weight:
sodium cyanide	50
sodium carbonate	50

Second composition:	Per cent by weight:
Potassium ferrocyanide	90
Sodium hydroxide	10

Cyanide baths with the second composition are less active, but at the same time they are less poisonous and, consequently, less dangerous in handling.

Care should be taken to select the cyaniding temperature within less than 10 to 15° C of the tempering temperature, i. e., 540 to 550°C. The holding time of cyaniding various tools is within the range of 10 to 25 minutes, depending upon their sections and the cyanide bath concentration. This process results in a cyanided case 0.01 to 0.02 mm thick.

After cyaniding tools are allowed to cool slowly in the still air. To ensure a uniform cooling, in order to prevent distortion of the articles, tools are suspended for cooling (e. g., in a net).

Immediately after cyaniding the articles treated must be neutralised and thoroughly rinsed, as is always required. All the safety rules discussed in detail in Chapter X must be strictly observed also in the liquid cyaniding process.

The *gas cyaniding*, or carbonitriding, of cutting tools is performed in the same manner as gas cyaniding at elevated temperatures. The gas mixtures used are prepared as follows: 1) 30 to 40 per cent by volume of ammonia, the rest being composed of pyrolysis, natural, or oil gases, and 2) 15 to 25 per cent by volume of ammonia, the remainder being composed of producer or town gas. Low-temperature gas cyaniding of cutting tools is conducted at the same temperature as liquid cyaniding. The gas cyaniding takes about 1 to 2.5 hours.

Solid (or dry) cyaniding resembles very much the pack carburising process. The tool to be cyanided by this method is packed in a closed container with a suitable cyaniding compound that may be of the following composition (in per cent by weight):

Charcoal	50
Sodium carbonate	20
Potassium ferrocyanide or potassium ferri- cyanide	30

The charcoal is generally crushed to pea size, dried and screened to remove the dust in just the same manner as in pack carburising. All the ingredients of the cyaniding compound are thoroughly mixed and dried at a temperature between 100 to 150°C. Packing the tools to be cyanided in containers with cyaniding compounds is performed in the same way as is usually done in pack carburising. The containers or pots are placed in a furnace after their covers have been thoroughly luted with a fire clay. The solid cyaniding process is accom-

plished at the same temperature as that of the liquid and gas cyaniding processes. After the containers are heated to a given temperature, the cyaniding process continues for a period ranging from 1 to 4 hours, depending upon the section of the tool being treated. When the cyaniding is completed, the tools are neutralised and thoroughly rinsed.

All three described methods of low-temperature cyaniding cutting tools of high-speed steels result in the same increase in the tool life, i. e., generally by 1.5 to 2 times. But they differ very much from each other in the procedure. The solid, or pack, cyaniding is the simplest of the three to carry out, but this method can be used only in small scale production. In large scale production either the liquid or gas cyaniding methods are employed, as both are more productive and more refined technically. Liquid cyaniding is apt to produce very good results, being a very rapid (taking considerably less time than other cyaniding methods) and more efficient process; but it may prove to be dangerous to a certain degree if the pertaining safety rules are not strictly observed. The gas cyaniding method is the most perfect, being simple in its performance, not involving the use of poisonous substances, cheap, and suitable for mass production.

It is undoubtedly advisable to apply the cyaniding process to tools that are: 1) worn away on the back edge but reground on the front face (such as thread milling cutters and gear cutters, and broaches); 2) worn away on back edges and pilots (additional cutting edges acting to guide the tool), but reground on main back edges (such as drills, counterbores). Tools that are worn away and reground on the back edge, such as screw taps, reamers, cylindrical and shank milling cutters, as well as tools that are worn away on their back edges and reground both on them and on the front face (e. g., grooving milling cutters), must be cyanided after each regrinding operation. Some tool-makers and heat-treaters doubt whether it is advisable to cyanide such tools. For cutting tools cyaniding is certainly not recommended.

The *sub-zero treatment* of high-speed steel cutting tools to increase their service life is fully justified first of all in those cases where the tools have to be tempered in three operations, because one operation of sub-zero cooling can be substituted for two tempering operations, thus reducing the over-all time required for heat-treating. The sub-zero treatment is recommended for incorporation in the heat-treating process whenever the tool is specified for tempering in one or two operations. The amount of the residual austenite in the steel structure is considerably increased by even a slight overheating (by 10 to 20° C) of the tool in hardening it, which in practice cannot be avoided in production conditions. This residual austenite can be reduced only by tempering the tools in three or even four consecutive

operations (the author of this book experienced a case when the desired hardness was obtained only after seven consecutive tempering operations). Yet the sub-zero treatment is apt to diminish the quantity of residual austenite, even if this amount is well above the normal level. It may prove of little use to apply the sub-zero treatment only as a means of correcting overheating (which usually becomes clear after the tempering operations have been performed), in order to "improve", or increase hardness up to the desired value, since the residual austenite tends to become stabilised, becoming more stable proportionally with the time elapsed since the hardening operation was completed. Therefore it is advisable to incorporate the sub-zero treatment in the heat-treating process applied to tools of high-speed steels.

The following sequence of operations in heat-treating tools of high-speed steel may be considered normal:

- 1) hardening;
- 2) sub-zero treatment;
- 3) ordinary tempering operation.

In such a sequence of heat-treating operations conditions will be created which are less favourable to the stabilisation of the residual austenite, than when the sub-zero treatment is performed after tempering.

In some instances, however, a different procedure has to be adopted, i. e., hardening is followed by tempering, then the sub-zero treatment is performed, and finally a secondary tempering operation. Such heat-treating procedure has to be applied to intricately shaped tools that are liable to develop high internal quenching stresses. If such tools were sub-zero-treated immediately after hardening, quenching cracks would be likely to develop. In recent times the service life of cutting tools of high-speed steels has begun to be increased by a new method, called *sulphidising*. This method consists of heating tools at a temperature between 550 to 560° C in a liquid or solid medium containing sulphuric compounds, such as iron sulphide FeS (usually in a solid condition), potassium rhodanate KCNS and sodium sulphite Na₂SO₃ (in a liquid state). In this process the surface layer of a tool is made to absorb the sulphides. The increased life of a tool with a sulphidised case is probably explained by the fact that the sulphides are apt to decrease the coefficient of friction between the tool and the part machined and chip forming, which results in the diminished heating of the cutting tool.

82. HEAT-TREATMENT OF MEASURING INSTRUMENTS

The main technical requirements that should be met by measuring instruments, being closely connected with the very nature of the service of these instruments, are as follows: first, the working parts of instruments should be resistant to wear, and, secondly, all instruments *must not change their dimensions* with time. In service when taking measurements the working parts of measuring instruments, particularly of templates and snap gauges, are subject to wear, resulting in a gradual change in their dimensions. For that reason the measuring instruments should be very resistant to wear. To increase the wear-resistance of these instruments they must be manufactured of high-carbon steels, and also hardened for very high hardness. But steel in as-hardened condition, possessing a high hardness, is liable to ageing. Therefore the heat-treating process applied to measuring instruments must ensure stable dimensions that are not likely to change with the time.

With regard to these two main requirements, measuring instruments are manufactured of steels that are capable of obtaining a high hardness (Rockwell C 50 to 60) through heat-treatment and, consequently, high wear-resistance. The heat-treating process is designed to give the instruments both a high hardness and stable dimensions.

Templates, snap gauges and other measuring instruments are manufactured of the following steels:

Templates and gauge blocks of high-accuracy classes . . .	X, XΓ
Ditto, of long and intricate shape	XBF
Ditto, of simple shape and low-accuracy classes . . .	Y10A, Y12A
Flat templates and snap gauges	15, 20 or 50, 55

High-precision measuring instruments are made of alloy steels, since alloy steel measuring tools are more stable in dimensions. This is explained by the fact that measuring instruments from alloy steels are quenched in oil, which results in lower internal stresses and thus in less warping (or plastic deformation) because of the stress-relieving.

The heat-treating process, applied to templates and other measuring instruments made from carbon steels of grades Y10A and Y12A, and alloy steels of grades X, XΓ and XBF, consists of the following operations:

- 1) stock annealing;
- 2) rough machining with allowance of 0.5 mm for finish-machining;
- 3) preliminary quenching in water (for carbon steels) or in oil (for alloy steels);

- 4) preliminary tempering at a temperature between 650 to 670° C;
- 5) finish-machining with allowance left for grinding;
- 6) hardening by preheating to 550 to 600° C in a batch-type furnace and final heating in a salt bath maintained at the proper hardening temperature; quenching in water with transferring to the oil bath (for tools of carbon steels), or in oil (for tools of alloy steels); the oil temperature must be about 50° C;
- 7) degreasing;
- 8) cooling to a temperature of -10° C (or even lower, though in most cases this is not necessary) or, if there are no refrigerating units available in the heat-treating shop dipping into fresh running water (the lower its temperature the better);
- 9) long tempering in an oil bath at a temperature between 170 to 190° C (or 190 to 210° C for tools of the XB grade steel) for a period of time as follows:

Diameter or thickness, mm	Time of tempering, hours
up to 10	4 to 5
10 to 20	6 to 8
20 to 30	8 to 12
30 to 50	12 to 15

- 10) inspection;
- 11) preliminary grinding;
- 12) second tempering at a temperature between 150 to 160° C for a period of 1 to 2 hours to remove the internal stresses set up in the grinding operation; this second tempering is applied only to high-precision instruments;
- 13) final grinding.

The preliminary quenching and the first high-temperature tempering are performed in order to prepare the steel structure for final hardening; this is brought about by refining the steel and improving its hardenability.

The manufacture of flat templates and snap gauges of carburising steels (grades 15, 20) differs from the production process considered above mainly in that after rough machining the instruments are carburised to a depth of 1 to 2 mm (depending upon the kind of instrument and the allowance specified). After carburising they are subjected to a preliminary quenching in water from a temperature between 840 to 860° C or normalised. Final quenching in water is performed from a temperature of 770 to 790° C, followed by tempering at a temperature between 150 to 170° C for a period of 2 to 3 hours.

Flat templates and snap gauges may also be manufactured from medium-carbon steels of grades 50 to 55; in such cases the instruments are subsequently hardened by applying high-frequency electric currents.

83. HEAT-TREATMENT OF HAMMER DIES AND DIE MOULDS

Hammer dies for drop-forging work under severe conditions. First, they are subjected to high dynamic stresses set up by intensive forging strokes continuously applied to the stock. Secondly, they are alternately heated by the heat of the stock being forged and cooled either by mazout or a solution of sodium chloride. Therefore the steels for hammer dies should possess high hardness, strength and toughness not only at normal (room) temperatures, but at the elevated temperatures as well (say, 400 to 600° C). To secure high mechanical properties throughout the whole body of a die (hammer dies generally have large dimensions), the steel must possess *good hardenability*. Its *thermal conductivity* should be also *fairly high* to prevent the setting up during service of a considerable temperature differential in various parts of the hammer die, since this is conducive to the development of high internal stresses. Finally, steels for hammer dies must be heat-resistant, i. e., stable to tempering, in order to retain their hardness and strength while heated in service.

Steels for hammer dies must therefore conform to very high requirements. Plain carbon and low-alloy steels are not suitable for hammer dies. Only dies of simple shape and small dimensions may be manufactured of V7A grade carbon steel or 7X3 grade alloy steel.

All large and intricately shaped hammer dies designed for long service must be made from the complexly alloyed steels.

The most suitable for hammer dies and best meeting the above-mentioned requirements are the steels of following grades: 5XHM, 5XGM, 5XBI', 5XB2C, 5XHT, 5XHB, 5XCH, and 5XH2CBΦ. All these steels are characterised by a low carbon content (0.5 per cent approximately) determining their high impact strength, as well as by the fact that they are complexly alloyed, which gives them high hardenability, increased tensile strength, and high stability while hot (heat resistance, or resistance to tempering).

These steels, as well as the steels of grades 5XBI', 5XB2C, and 5XH2CBΦ, are recommended for use only for especially heavy-duty dies in technically well justified cases. Hammer dies must primarily be made from steels of grades 5XHT, 5XHB, and 5XHC.

Hammer dies are generally heat-treated after a preliminary machining, in which operation they are given the desired shape. In this preliminary machining a small allowance is made to be machined after the heat-treatment. The practice of dividing the machining of hammer dies into two stages, consisting of preliminary and finishing operations, is aimed at facilitating the machining, on the one hand, and at obtaining the exact dimensions, which may be otherwise distorted in heat-treating, and a clean surface, on the other hand.

Hammer dies of complexly alloyed steels are heat-treated as follows:

Heat-Treating Processes for Hammer Dies

Table 27

Name of operation	Time of operations in hours and minutes for hammer dies of a given height in mm								
	250	300	350	400	450	500	550	600	700
<i>Hardening</i>									
Holding after charging	0-30	0-30	0-30	1-00	1-00	1-30	1-30	2-00	2-00
Heating to temperature of hardening	7-00	8-30	10-00	11-00	12-30	13-30	15-00	16-00	19-00
Holding time at hardening temperature	1-30	1-50	2-10	2-30	3-00	3-00	3-20	3-40	4-10
Precooling in open air	0-10	0-13	0-15	0-17	0-23	0-25	0-30	0-35	0-40
Cooling in oil	0-30	0-40	0-50	1-00	1-10	1-20	1-30	1-40	2-00
<i>Tempering</i>									
Heating to temperature of tempering	7-30	9-00	10-30	12-00	13-30	15-00	16-30	18-00	21-00
Holding time at tempering temperature	1-30	1-50	2-10	2-30	2-40	3-00	3-20	3-40	4-10

Note: Times of heating and holding are indicated for oil- and gas-fired furnaces. For electric furnaces the indicated times are to be increased by 20 per cent.

1) charging in a furnace heated up to 600° C and holding there for a period of time, as indicated in Table 27 (according to data of G. L. Livshits and E. V. Smirnova);

2) heating in a furnace to the proper hardening temperature at a rate of 25 to 35° C per hour (for dies 250 to 400 mm high) or 15 to 25° C per hour (for dies 450 to 700 mm high); Table 27 indicates the heating time; the hardening temperatures are selected as follows:

Grade of steel	Hardening temperature, ° C
5XHМ	820 to 860
5XГМ	850 to 870
5ХВГ	850 to 900
5ХВ2С	860 to 880
5ХТ	830 to 850
5ХНВ	840 to 860
5ХНС	850 to 870
5ХН2СВΦ	870 to 890

3) holding times at the hardening temperature are to be selected from Table 27;

4) hammer dies are cooled in oil; before being immersed in the oil the die is precooled in air to a temperature between 750 to 700° C, the approximate time of precooling in the open air being indicated in Table 27; hammer dies are cooled in oil to a temperature between 200 to 150° C (the oil is slightly fuming on the surface of the die at that temperature), the approximate time of maintaining the die in the oil bath being also indicated in Table 27;

5) immediately upon hardening, placing in a tempering furnace heated to a temperature between 300 to 350° C;

6) heating to the tempering temperature for the period of time indicated in Table 27, the tempering temperatures being equal (in ° C) to:

Small hammer dies	525 to 550
Medium-size hammer dies . .	550 to 575
Large hammer dies	575 to 600

For hammer dies made from 5XHC grade steel a tempering temperature is adopted at 25° C above the indicated values, while for those of 5XHT grade steel, within less than 50° C of these values;

7) holding time at a tempering temperature is to be taken from Table 27;

8) shanks of dies up to 400 mm high are tempered at a temperature between 590 to 610° C, and those of larger dies at 620 to 640° C. Shanks should be tempered as indicated in Fig. 106; they may be also tempered on a base plate heated to between 700 to 720° C.

Working parts of hammer dies in as-treated condition must have the following hardness:

	Diameter of Impression, mm	Brinell hardness number H_B
Small dies	3.1 to 2.9	388 to 444
Medium dies	3.3 to 3.1	341 to 388
Large dies	3.5 to 3.3	302 to 341

The diameter of indentation of a steel ball on the shank of a hammer die must be by 0.3 to 0.5 mm greater than that produced on the working surface.

Mould dies for die-casting machines work under more severe conditions than the case may be with hammer dies. In the die-casting process, molten metal is forced under a high pressure (amounting to scores of atmospheres) into the impression of a mould in order to solidify there. The temperatures of molten alloys used in modern die-casting practice are as follows: zinc-base alloys, about 450° C; aluminium- and magnesium-base alloys, about 600° C; copper-base alloys, about 1000° C. The temperatures of the molten alloys to be

cast determine the steel grade to be used for making die moulds for die-casting machines. This may be represented as follows:

	Grade of steel
Copper-base alloys	3X2B8
Aluminium- and magnesium- base alloys	4X8B2, 4XB2C, 30XFC
Zinc-base alloys	30XFC, 40

Mould dies for die-casting machines are heat-treated in the same way as hammer dies. The hardening and tempering temperatures are selected as follows:

Grade of steel	Hardening temperature, °C	Tempering temperature, °C
3X2B8	1050 to 1100	600 to 620
4X8B2	1050 to 1100	575 to 600
4XB2C	950 to 1000	550 to 600
30XFC	890 to 920	520 to 560

84. HEAT-TREATMENT OF DIES FOR COLD-SWAGING AND COLD-HEADING PROCESSES

All dies used in *cold-swaging processes*—blanking, cut-off, bending, drawing dies—consist of two main parts: the punch and die, as well as of various accessories such as lift-outs, guide pins, bushings, etc. During use of the dies the most severe conditions exist for die and punch; they undergo high pressures, exerted by a swaging press, and during service they are always subjected to friction with the material to be swaged. For this reason *high wear-resistance* is the main requirement to be met by the steels used for making dies and punches.

Of all tool steels, the most wear-resistant are the high-chromium steels of the carbide class of the grades X12, X12M, X12Φ, and X12Φ1. The X12Φ1 grade steel is specified by the U.S.S.R. State Standard on the stock for dies; its chemical composition is as follows: 1.45 to 1.7 per cent carbon, 11.0 to 12.5 per cent chromium, and 0.7 to 0.9 per cent vanadium. The X12Φ grade steel differs from the X12Φ1 grade by its somewhat lower carbon content (1.40 to 1.60 per cent) and considerably lower vanadium content (0.2 to 0.4 per cent). In this group of steels, the X12 grade steel is the least good from the point of view of toughness, so the three other grades are always preferred in practical use.

The X12M grade steel, being alloyed with expensive and scarce molybdenum, should be used only for making heavy-duty dies in technically well justified cases.

In addition to a high resistance to wear, an outstanding characteristic of the high-chromium steels of the indicated four grades is that they have a *minimum change in dimensions* on hardening.

Besides the above-mentioned high-chromium steels of the carbide class, less heavily stressed cold-swaging dies may be made of carbon steels, the grades of which are classified as Y10 and Y11; or low-alloy steels of the pearlitic class of the following grades: X, X09, 9XC, XΓ, and XBF.

Two kinds of heat-treating process are applied to dies and punches of high-chromium steels of grades X12M, X12Φ, and X12Φ1, depending upon whether the heat-treatment is desired to secure permanence of the die parts or whether such a high resistance to deformation is not necessary, and the dies can be ground to size after heat-treatment.

The heat-treating process applied to drawing, cutoff, and blanking dies, which are not necessarily required to retain exact and permanent dimensions in heat-treating operation, consists of the following stages:

1) annealing the stock after forging, consisting of heating to a temperature between 850 to 870° C, holding there after heating through for a period of several hours (usually not less than 2 to 2½ hours), and cooling to a temperature between 450 to 500° C at a rate of 15 to 20° C per hour;

2) rough or preliminary machining;

3) preliminary quenching in oil from a temperature of about 1000° C;

4) preliminary high tempering (or drawing-back) at a temperature between 630 to 660° C;

5) finish-machining;

6) hardening, consisting of the following operations: heating to a temperature between 800 to 850° C for a period of time calculated on the basis of 1 to 2 minutes for each millimetre of the section thickness; final heating up to the proper temperature of hardening, as follows:

Grade of steel	Hardening temperature (° C) at which the final desired hardness may be obtained	
	54 to 57 Rockwell C	58 to 60 Rockwell C
X12M, X12Φ	1020 to 1030	1050 to 1060
X12Φ1	1070 to 1080	1120 to 1130

The piece is maintained at the hardening temperature for a period of time determined on the basis of ½ to 1 minute for each millimetre of the cross-section thickness. Parts of large dies (Rockwell C 54 to 57) are cooled in oil, and parts of small dies in the open air. Parts of dies required to have the Rockwell C hardness ranging from 58 to 60 are transferred from the hardening furnace into another furnace

heated to 450° C, or into a bath (as is recommended for large parts) heated up to the same temperature, and are held in the furnace for a period of 40 to 50 minutes and in the bath for a period of 30 to 40 minutes; after such a precooling, they are cooled in the open air;

7) tempering the parts that are to have the range of Rockwell C hardness from 54 to 57 at a temperature between 350 to 400° C for a period of 2 hours after they have been heated throughout; tempering of the parts that are required to have the range of hardness from 58 to 60 Rockwell C is performed at 520° C for a period of 2 hours, provided they are heated throughout;

8) grinding to final size.

The heat-treatment of intricately shaped dies and punches of blanking and bending dies that cannot be ground to size and should not, therefore, be subject to any changes in dimensions on heat-treating, differs from that just considered by the application of rather different hardening temperatures, the most essential difference being the tempering practice, which operation in the first case is intended to be the finish heat-treatment as well. The essentials of the finish heat-treatment and methods of its performing have been already discussed in detail in Chapter VII of this book.

Dies and punches of carbon and low-alloy steels are heat-treated by a considerably simpler process consisting either of normalising or (for large articles) of preliminary hardening in oil to obtain the fine-grain structure, followed by the final hardening and tempering operations. Temperatures of hardening and tempering (according to data of Y. G. Rauzin and Y. A. Geller) are indicated in Table 28.

Table 28

**Tempering of Hardening and Tempering Dies
and Punches of Cold-Heading Dies**

Grade of steel	Hardening temperature	Temperatures of tempering to hardness	
		Rockwell C 55 to 60	Rockwell C 50 to 55
Y10	790 to 800	250 to 270	275 to 325
Y11	790 to 810	250 to 270	275 to 325
X	835 to 845	200 to 260	270 to 330
X09	825 to 835	190 to 250	250 to 320
9XC	845 to 865	260 to 320	330 to 340
XΓ	835 to 845	200 to 250	260 to 340
XBF	825 to 845	230 to 280	280 to 360

Accessories to the dies, such as guide posts, bushings, pins, lift-outs, pushers, etc., are to be made only of carbon steels, both carbu-

rising and tool grades. They are heat-treated by the ordinary heat-treating processes.

Working parts of cold-heading dies, i. e., dies and punches, should combine high tensile and compressive strength and increased hardness with fairly high impact strength. Of carbon tool steels, these requirements are best met by the Y10A grade steel. Cold-heading dies and punches are sometimes also manufactured of 9XC grade steel, belonging to the class of alloy tool steels.

The heat-treating process applied to cold-heading dies and punches made of the Y10A grade steel consists of quenching in water from a temperature of between 820 to 830° C, transferring into oil and tempering at a temperature of between 230 to 260° C for a period of 2 to 5 hours. To increase the hardenability of dies and punches over 60 mm in diameter, they are quenched from a somewhat higher temperature, e. g., between 850 to 880° C. The range of hardness for cold-heading dies and punches in as-hardened condition should be Rockwell C 55 to 59.

PART FOUR

**HEAT-TREATMENT OF CAST IRONS
AND NON-FERROUS ALLOYS**

Chapter XVII

HEAT-TREATMENT OF IRON CASTINGS

85. CLASSIFICATION AND STRUCTURE OF CAST IRONS

Cast iron is an alloy of iron and carbon containing more than 2.0 per cent of carbon. Ordinary cast irons contain 2.5 to 4 per cent carbon. In addition to iron and carbon, commercial cast irons, like the steels, usually contain also silicon, manganese, phosphorus, and sulphur; alloy cast irons may also contain chromium, nickel, and aluminium.

Carbon, being the most essential (after iron) ingredient of any cast iron, may be present in its structure only in the form of *cementite* Fe_3C or only in the form of *graphite*, or, which is most common, simultaneously in the forms of cementite and graphite. Those cast irons in which all the carbon is in the combined form, i. e., in the form of cementite, are known as *white* cast irons. *Gray* cast irons are characterised by the presence of all or the most of carbon in the form of graphite, the lesser part (usually under 1 per cent) being in the form of cementite. Finally, those cast irons in which the greater part of carbon is present in the form of cementite, and the lesser part in the form of graphite are called *mottled* cast irons.

The machine-building industry uses mainly gray iron castings. Castings consisting throughout of white iron cannot be used, since they are excessively brittle and almost completely incapable of being machined. White iron castings provide the stock to be heat-treated to obtain castings with a structure of so-called *malleable* cast iron, in which the carbon appears wholly or mostly as graphite in a form known as temper carbon.

Castings with a chilled surface layer, the central portion or core consisting of the gray iron, are also used, e. g., for cast iron rolls for cold rolling. The chilled surface layer is characterised by a high hardness and increased wear-resistance, which is very important for these parts. Between the chilled case of white iron and the core of

gray iron there is an intermediate layer consisting of mottled iron.

Castings with a structure consisting wholly of mottled cast iron are not produced commercially; if for some reason such a structure is developed, the casting is usually either rejected or corrected by heat-treatment.

The most essential structural constituent of gray cast irons is the graphite (Fig. 154) which is distributed in the form of long, thin flakes that appear in the surface of a microsection as elongated and twisted inclusions.



Fig. 154. Photomicrograph of gray cast iron, showing graphite (the microsection is not etched), 300 $\times$.

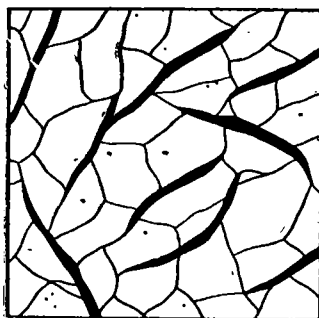


Fig. 155. Microstructure of ferritic gray iron, 300 $\times$.

The main metallic matrix of the gray cast iron (with the exception of the graphite) is composed of an alloy of iron and carbon. As has been stated above, the carbon content in the matrix of gray iron is usually about 1 per cent. Thus, with some approximation, the metallic matrix in the structure of gray iron may be regarded as a steel that follows the mode of behaviour of steels in general. Such an approach to the structure of gray cast irons allows the ready distinguishing of the possible kinds of gray iron structures, e. g.:

graphite + ferrite (Fig. 155);
graphite + ferrite + pearlite (Fig. 156);
graphite + pearlite (Fig. 157);
graphite + pearlite + cementite.

For the majority of gray iron castings the most typical structure is either composed of graphite, ferrite, and pearlite (as in Fig. 156), or of graphite and pearlite (as in Fig. 157). Cast iron with a structure of graphite and ferrite (Fig. 155) is very soft, weak, and possesses a low resistance to wear, whereas that with a structure of graphite, pearlite and cementite is very hard and difficult to machine.

The fact that the metallic matrix of the gray cast irons is considered as a "steel" permits the application to gray iron castings of the same kinds of heat-treatment as are applicable for steels, i. e., annealing, normalising, hardening, and tempering. A gray iron casting can in fact be hardened, the resulting structure being composed of graphite and martensite, and if it is drawn-back after hardening, its structure will contain not only graphite, but also sorbite.



Fig. 156. Microstructure of ferrito-pearlitic gray iron, 300 $\times$.

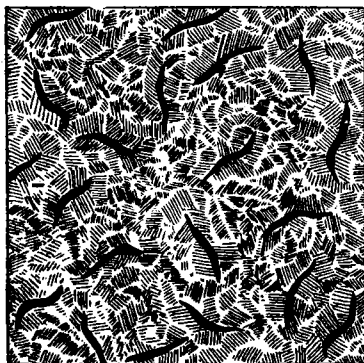


Fig. 157. Microstructure of pearlitic gray iron, 300 $\times$.

However, gray iron castings are only very seldom subjected to hardening with subsequent tempering, or even to normalising. This is explained by the fact that the structure of gray iron castings contains graphite, a weak and porous substance. The graphite flakes interrupt the continuity of the metallic matrix, so the strength of a gray cast iron largely depends upon their quantity and distribution, as well as on their size and shape, and to a less extent upon the properties of the metallic matrix. However high a hardness may be developed in the metallic matrix of a gray iron by means of a heat-treatment, it will be almost completely offset by the graphite present. This is why the strengthening heat-treatment is hardly ever applied to gray iron castings. This is also the reason why the practice of alloying gray cast irons to improve their mechanical strength has hardly come into use. The gray cast irons are alloyed only to improve their physical and physico-chemical properties. Alloyed acid-proof, heat-resisting, and non-magnetic cast irons are available.

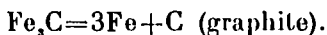
The strength of gray cast irons may be increased by another method, i. e., through changing the quantity of the graphite and especially by changing its size and shape. Methods of obtaining malleable, ductile, and high-strength cast irons have been developed on that very principle. These methods will be discussed below.

86. HEAT-TREATMENT OF GRAY IRON CASTINGS

At machine-building plants gray iron castings *are* most frequently *annealed to improve their machinability*. Gray iron castings that have for some reason acquired a chilled surface with a structure of white iron must definitely be annealed. The chilled case may be caused in several ways: by a wrong chemical composition of the gray cast iron (most often an insufficient silicon content for a given section thickness of the casting), by an excessive rate of solidification because the moulding sand was too humid, etc. Gray iron castings cast in gravity-type permanent moulds must be annealed in nearly all cases. An increased hardness may be obtained also in those gray iron castings that have no chilled case, but have been rapidly cooled after solidification; e. g., when they have been knocked out too early from moulds (being still red hot) and have been cooled not in moulds, but in the open air. Although such castings have a gray iron structure, their high hardness is explained by the presence of the secondary cementite, on the one hand, and the high degree of dispersion of the ferrito-cementite conglomerate, on the other hand. The structure of such cast irons consists of graphite, sorbite and cementite.

To decrease the hardness of gray iron castings and improve their machinability it is necessary to anneal them by means of heating to a temperature of between 850 to 950° C and holding at that temperature for a period of 1 to 2 hours. The annealed castings may be in most cases cooled in the open air, their hardness nevertheless considerably decreasing.

When gray iron castings are annealed the cementite decomposes according to the formula:



The higher the heating temperature the more intensively the graphitisation of the cementite progresses. Small castings of simple design may be heated in a salt bath to a temperature of 1050 to 1150° C. At this temperature only a very short holding time is needed for the cementite to decompose and the hardness to be decreased to 150 to 130 H_B .

In annealing gray iron castings to decrease their hardness it should be taken into account that castings of thin sections are liable to warping, the degree of deformation increasing with the annealing temperature. Castings that are liable to warping have to be annealed at lower temperatures, say 800 to 850° C, with the holding time of heating considerably lengthened (2 to 5 hours); in such cases the cooling of the castings should be conducted in the furnace to a temperature of 400 to 500° C. In some instances this is the underlying reason why the annealing operation has to be omitted entirely.

The annealing of gray iron castings as a means for decreasing their hardness should not be misused. In point of fact, the annealing not only decreases the hardness of a cast iron, but is also conducive to a decrease in its strength. Therefore if gray iron castings are required to meet more exacting requirements of strength (e. g., castings from gray irons of grades CY 21-40, CY 24-44, CY 28-48, etc.), annealing may prove to be injurious or even completely impracticable. The advisability of annealing in such cases should be decided after adequate consultation with a design engineer. Also possible, though it is considered undesirable, is an annealing of gray iron castings followed by hardening and drawing-back after machining.

Hardening with subsequent tempering may be applied to some gray iron castings requiring a *high wear-resistance*, e. g., rollers, bushings, chain wheels for combines, eccentrics, piston rings, etc. The castings are heated for hardening up to a temperature ranging between 800 to 900° C, the lower limit being selected for intricately shaped articles and the higher one for articles of simple design. The cooling is also performed differently: castings of simple shape may be quenched in water, and those of more intricate design in oil. The hardened castings are tempered at a temperature between 350 to 450° C for a period of 1 hour. The range of hardness produced by such a heat-treating process is 300 to 350 H_R .

Gray iron castings may be made more resistant to wear also through an isothermal quenching procedure from a temperature of between 830 to 870° C in a saltpeter bath maintained at 280 to 350° C. The castings are held in the saltpeter bath for a period of 30 to 60 minutes, depending on their weight.

Piston rings are subjected to a special heat-treatment called *thermal* or *hot-fixing*. This operation consists of fitting the piston rings on a mandrel of diameter somewhat greater than the internal diameter of the ring, and heating the whole assembly in a furnace to a temperature between 550 and 600° C for a period of 1 to 2 hours. As a result of such treatment the piston rings increase in their diameter (their cut length increases), with the result that they acquire an excellent resilience, which is precisely what is desired. In service piston rings always tend to spring, thus producing pressure against the cylinder wall. If after hot-fixing a piston ring is shorter than the desired cut length it must be subjected to this heat-treating operation the second time, usually at a somewhat higher temperature (by 20 to 30° C).

Gray iron castings are frequently subjected to *stress-relieving annealing*, or *artificial ageing*. Many castings, particularly those of uneven section, have considerable internal stresses, usually being set up by non-uniform cooling. Gradually, over a long period

of time, these internal stresses tend to be relieved spontaneously. This spontaneous relieving of internal elastic stresses is accompanied by some plastic deformation, with the result that a casting changes its shape and dimensions to some extent. For this reason heavy-duty gray iron castings that must not change their shape and, consequently, their dimensions with time (e. g., machine-tool beds), were formerly produced months ahead of the intended machining date in order that they might "season". However, such seasoning or natural ageing, which may be regarded as a low-temperature annealing for a very long time, not only prolongs very much the production cycle but is also insufficiently effective, e. g., it takes from 6 months to 1 year to relieve only 30 to 40 per cent of the initial internal stresses. At the present time, therefore, an artificial, or quick, ageing is performed. This involves charging gray iron castings in a furnace being either cold or heated up to a temperature between 200 to 250° C, and slow heating there (at a rate of 100 to 150° C per hour) to a temperature between 500 and 600° C. At this temperature the castings are maintained for a period of 3 to 5 hours, after which they are slowly cooled (at a rate of 25 to 75° C per hour) to a temperature of between 150 and 250° C. The diagram in Fig. 158 illustrates the effect of the temperature of artificial ageing upon the degree of removal of internal stresses.

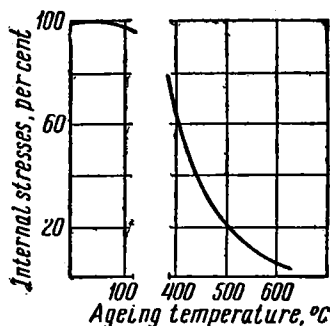


Fig. 158. Diagram showing the effect of temperature of artificial ageing of gray cast-iron castings upon the degree of stress relieving.

87. HEAT-TREATMENT OF WHITE IRON CASTINGS (PRODUCTION OF MALLEABLE IRON CASTINGS)

The insufficient strength of gray cast irons and their very low ductility and toughness are caused not so much by the comparatively large amount of graphite in their structure as by the clearly unfavourable size and distribution of this graphite in the form of long and thin flakes. Malleable cast-iron has all its graphite in the form of nodules, and its mechanical properties are considerably higher than those of gray cast iron. Let us compare the mechanical properties of gray and malleable cast irons having the same structure in terms of ferrite and graphite (the latter in the form of flakes and nodules).

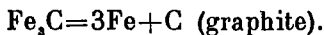
Gray cast iron	Malleable cast iron	
Tensile strength, kg/sq mm	10 to 12	25 to 30
Elongation, per cent	almost zero	15 to 20
Impact strength, kgm/sq cm	0.5	2 to 4

Malleable cast irons with a structure consisting of ferrite, pearlite and graphite, are even stronger (tensile strength up to 40 to 50 kg/sq mm), and though their ductility and toughness somewhat decrease as the quantity of pearlite increases, they remain considerably superior to those of gray cast irons.

Malleable cast-iron castings are produced in two main manufacturing stages: 1) production of white cast-iron castings, and 2) high-temperature annealing of these castings. The usual term "heat-treatment of malleable cast iron" is not altogether correct, since the heat-treatment is applied not to malleable cast iron but to the white cast iron, the former being obtained as the result of this heat-treating.

White cast iron, the castings of which are to be annealed to obtain malleable cast iron, should have as low a carbon content as possible (2.5 to 2.8 per cent carbon), because the lower the carbon content the less amount of graphite will form, thus causing the mechanical properties to be higher. However, the carbon content cannot be decreased too drastically, since this would very much complicate the melting process and markedly deteriorate the fluidity of the hot metal. The malleable iron must be also low in silicon and manganese, the former not above 1 per cent and the latter up to 0.5 per cent. With a higher silicon content a wholly or partly gray cast iron will be obtained, with the result that it will become impossible to convert it into malleable cast iron, for the graphite flakes cannot be transformed into nodules by any available means. Increasing the manganese content is desirable to facilitate the production of white cast iron; from the standpoint of the following annealing process, however, a high manganese content is highly undesirable, since it tends to stabilise the carbides.

White cast-iron castings are annealed in two stages to convert them into the malleable cast iron. The first stage of a typical heat-treating cycle consists of heating the castings at a temperature between 900 to 950° C, and the second stage of heating them at a temperature of between 760 to 720° C (Fig. 159). Before the annealing operation the structure of a white iron consists of two phases—*austenite* and *cementite* (the latter being a constituent of the *ledeburite* eutectic). The annealing does not affect the *austenite*; the *cementite* decomposes according to the following equation:



After the first stage of graphitisation the structure of a malleable cast iron consists of *austenite* and *graphite*, i. e., in the course of the

first annealing operation the white cast iron is converted into malleable cast iron. The carbon content in the austenite determined from a phase diagram is about 1 per cent at 900°C . If malleable iron is cooled after the first stage of graphitisation, its structure will then consist of graphite, pearlite and the secondary cementite or ferrite, depending upon the cooling rate (Fig. 160). The presence of cementite is completely inadmissible, and pearlite, too, is considered undesirable in most cases. The second stage of annealing is intended to decompose both the secondary and pearlite cement-

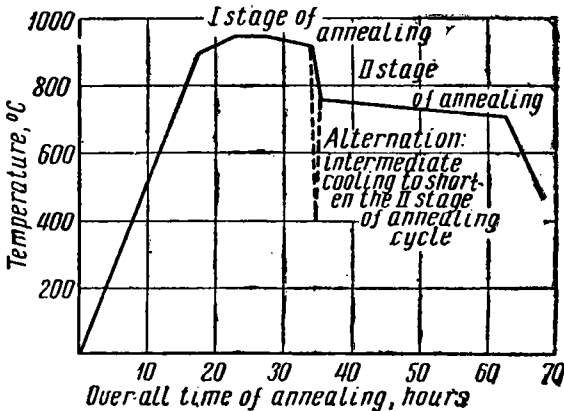


Fig. 159. Diagram showing the process of annealing white cast iron to produce malleable iron.

ites, i. e., to convert all the combined carbon of the free cementite and pearlite of the original white cast iron to a form of carbon which is nodular rather than flaky.

The second stage of annealing is performed at the pearlitic transformation temperatures; in this operation is formed the ferrite-cementite conglomerate, with the cementite of this mixture simultaneously decomposing. As a result of annealing the structure of malleable cast iron becomes similar to that shown in photomicrograph of Fig. 161.

The time for annealing a white cast iron so as to obtain a malleable cast iron is very long, continuing for three days and even longer in some cases. This is a very great disadvantage of malleable cast iron. Therefore it seems natural that the overall amount of castings of malleable cast iron, even though it is a material excellent in all respects, should not exceed a mere 2 per cent of the grand total of all iron castings used in the industry.

Several methods have been developed for shortening the time

required for heat-treating white cast-iron castings. One of these methods consists in *increasing the temperature of the first annealing operation*, e. g., up to 1000°C , since the rate of cementite decomposition accelerates considerably at that temperature. This is a perfectly sound method. However, if one considers not only the process of graphitisation, but the quality of the castings produced as well, then such an accelerated annealing may be resorted to only in a comparatively limited number of cases, because the castings become mechanically weaker and can be more easily deformed as the annealing temperature increases.

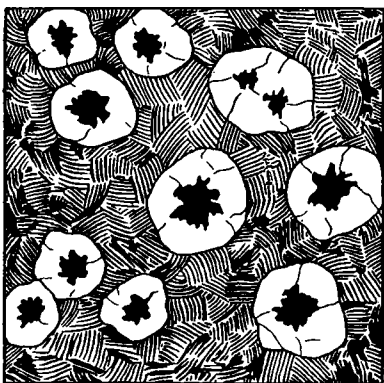


Fig. 160. Microstructure of ferrito-pearlitic malleable iron, which has been subjected only to the first stage of the annealing cycle, $300\times$.

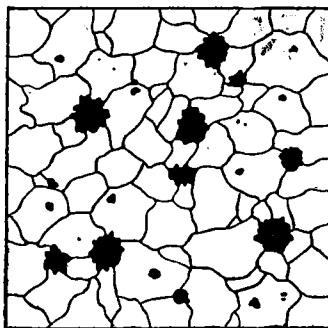


Fig. 161. Microstructure of malleable iron, showing graphite and ferrite, $300\times$.

The second method of shortening the heat-treating cycle consists of *hardening* the white cast-iron castings before annealing in order to set up artificially the *internal stresses* in them. In point of fact it has been found by some metallographers that the internal stresses present in the white cast-iron structure favour the process of graphitising cementite. Evidently this method of heat-treating may be applied only to castings of relatively simple design, in which the hardening operation does not set up excessive internal stresses leading to the development of quenching cracks.

Finally, the heat-treating cycle may be shortened also by *preliminary forming of nuclei of graphitisation*. This is obtained through a special treatment of the liquid cast iron when it is tapped from the cupola; it involves nodulising of the graphite and is generally called *modification*. However, even the application of this method is connected with some difficulties in practice, as the graphitising nuc-

lei formed by modifying the liquid cast iron may cause the process of graphitisation to begin during the primary crystallisation of castings, and this is highly undesirable. In such a case the modification process may prove to be detrimental rather than useful, for it will only deteriorate the physical properties of cast iron. *

There is no doubt that more refined and efficient methods of speeding up the process of annealing white cast iron will continue to be sought.

88. HEAT-TREATMENT OF HIGH-STRENGTH IRON CASTINGS

In modern practice the strength properties of gray cast iron are improved through a special treatment of the liquid cast iron, called modification. This involves adding to the liquid cast iron, before it is poured into moulds, special finely crushed substances, called modifiers or inoculators, capable of producing a great number of nuclei of graphitisation in it. At the present time two types of inoculants are used, of which the first (ferrosilicon, silicocalcium, etc.) produces a great quantity of nuclei of graphitisation without simultaneously changing the form of the graphite being produced. The graphite produced by the inoculants of this group is in the form of fine flakes (see Fig. 157). Accordingly, modification of cast irons by the addition of ferrosilicon and other similar inoculants results only in high strength properties, while ductility and toughness remain as low as without this treatment.

Inoculants of the second type, such as magnesium and magnesium alloys, also affect the shape of the crystallising graphite, which is, produced in the form of nodules or spheroids. The structure of a cast iron modified by the addition of magnesium only contains a great amount of cementite. For this reason the treatment with magnesium is followed by additional modifying by ferrosilicon. In some instances the modification procedure comprises treating the cast iron simultaneously with magnesium and ferrosilicon in one operation.

Cast irons modified by the addition of magnesium and ferrosilicon are called high-strength ductile irons. The structure of a ductile iron is shown in the photomicrograph in Fig. 162. Ductile iron responds

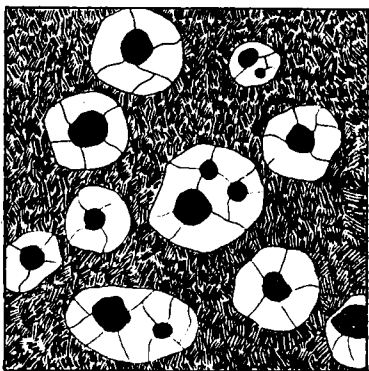


Fig. 162. Microstructure of ductile iron after modification by additions of magnesium and ferrosilicon, 300 $\times$.

to many types of heat-treating processes of which annealing is the most important. Although annealing somewhat decreases the strength properties, at the same time it markedly increases its ductility and toughness, which become comparable to those of cast steel, the elongation being not less than 10 per cent and the impact strength in excess of 3 kgm/sq cm.

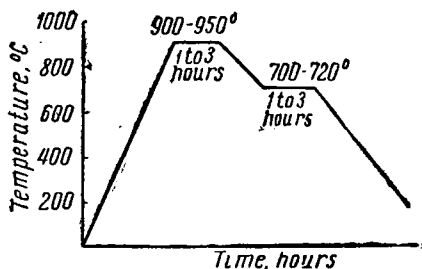


Fig. 163. Diagram showing the annealing process for ductile iron castings.

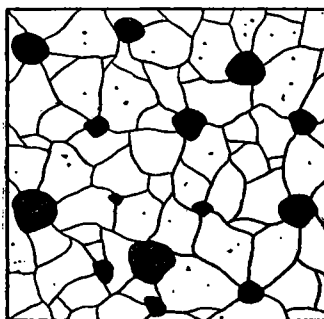


Fig. 164. Microstructure of high-strength ductile iron in as-annealed condition, showing graphite and ferrite, 300 $\times$.

The annealing of ductile cast-iron castings is a very simple process that takes a relatively short time as compared with the long annealing of a white cast iron to produce malleable iron. The diagram of Fig. 163 illustrates the annealing of ductile iron castings. The structure of a high-strength ductile iron in as-annealed condition consists of graphite and ferrite (Fig. 164).

Chapter XVIII *

HEAT-TREATMENT OF COPPER AND COPPER-BASE ALLOYS

89. HEAT-TREATMENT OF COPPER

Copper, like other pure metals, can only be heat-treated in one way, i. e., annealed *to recrystallise the metal*. According to the well-known formula of A. A. Bochvar, pure copper recrystallises at 270° C. However, at that temperature the rate of recrystallisation is slow, so that in practice pure copper is annealed for recrystallisation at considerably higher temperatures, e. g., 500 to 700° C. Further increase in the annealing temperature is not suitable when annealing is

conducted in the atmosphere of air as in this case the copper is intensively oxidised at temperatures above 700° C (Fig. 165).

After recrystallisation annealing copper must be quenched in cold water, for at such a rapid cooling the scale, formed on its surface will completely flake off, leaving an absolutely clean surface of pure (or red) copper. Other methods of removing the scale (e. g., through pickling) are considerably more complicated, becoming frequently even impracticable when there is a fairly thick layer of scale on the surface of the metal.

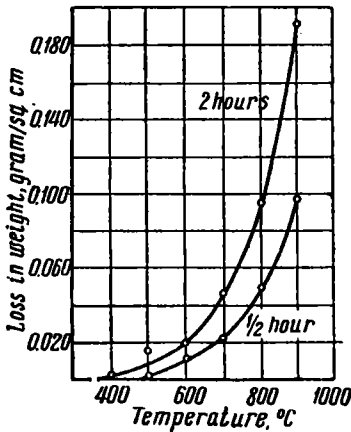


Fig. 165. Diagram showing the effect of the annealing temperature upon the oxidation of copper.

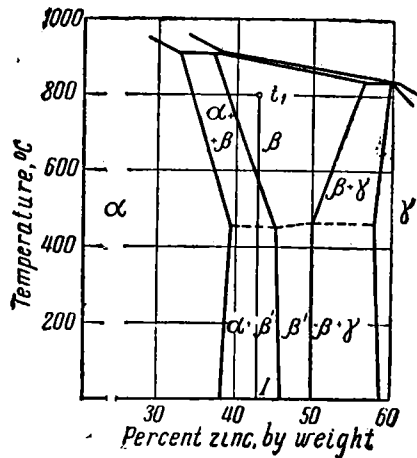


Fig. 166. Portion of phase diagram for brasses.

90. HEAT-TREATMENT OF BRASSES

Like pure copper, single-phase alpha brasses (brasses of grades up to J162) may be subjected only to the *recrystallisation annealing*, which is conducted at a temperature between 700 to 730° C. Double-phase brasses with a structure consisting of a mixture of two solid solutions, alpha and beta, may be either annealed to recrystallise the metal or subjected to such types of heat-treatment as *hardening*. Such a heat-treatment is made possible by the fact that double-phase brasses have phase transformations (Fig. 166). The diagram in Fig. 167 illustrates the change in hardness during tempering of a brass of the grade J1C59-1 hardened from a temperature of 850° C.

A *low-temperature annealing* may be given to articles of cold-worked brass to prevent cracks from developing spontaneously.

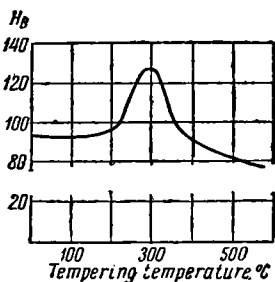


Fig. 167. Diagram showing the change in Brinell hardness in tempering of grade JIC59-1 double-phase brass quenched from 850° C.

Some cold-worked copper-base alloys are subject to develop cracks quite spontaneously without any external influence whatsoever; e. g., they tend to crack when being stored in a storehouse. This tendency is especially pronounced in brasses which have been cold-worked, e. g., brass caps, cartridge cases for rifles, caps for incandescent lamps, pipes for air-coolers, etc. It has been found that the spontaneous cracking occurs in articles in which the internal stresses (of the tension type) are accompanied by corrosive conditions. Even a very weak corrosive environment, e. g., the ordinary atmosphere of industrial centres containing some amounts of ammonia and sulphur dioxide, is capable of causing the brass to crack

at the grain boundaries, such cracking being known as season cracking.

The reason why it is so difficult to reduce the tendency of brasses to season cracking by means of annealing is that the annealing temperatures (above 300° C) at which the internal stresses can be completely removed are well above the recrystallisation temperature for brass (250 to 300° C), thus causing the hardness of a brass to be somewhat lowered. Nevertheless, this has to be accepted as the least of two evils. So brass articles are annealed at 300° C for a period of 1 hour in order to relieve the internal stresses and thus reduce the tendency to season cracking. After annealing the articles may be cooled at any rate.

91. HEAT-TREATMENT OF BRONZES

Most bronzes used in the machine-building industry are not subjected to any heat-treatment except annealing for recrystallisation. Only one of the bronzes, beryllium bronze (or beryllium copper), is used exclusively in the heat-treated condition. The heat-treatment of beryllium bronze is made possible by changing the solubility of beryllium in its solid solution with copper (Fig. 168).

Two grades of beryllium bronzes are available, Bp. B2 and Bp. B2.5, containing 2 or 2.5 per cent beryllium respectively, as it is clear from the grade designations. The heat-treatment of beryllium bronzes consists of quenching them in water from a temperature between 760 and 780° C, followed by tempering (which is usually called ageing or refining) at a temperature between 310 and 330° C for a period of

2 to 2½ hours. As a result of such a process the beryllium bronze markedly increases in hardness, the range of Vickers hardness being 320 to 400 in as-treated condition. The heat-treatment results simultaneously in a marked increase in the tensile strength (up to 150 kg/sq mm) and yield point of the metal.

This determines the uses of beryllium bronzes in the industry. They are chiefly employed in machine-building and especially in the manufacturing of tools and instruments, where they are used for spring contacts exposed in service to high temperatures and corrosive conditions. Beryllium bronzes, being copper-base alloys, possess a fairly high electrical conductivity and high resistance to corrosion, and acquire a high elasticity after heat-treatment. There is no other copper-base alloy used on a commercial scale which combines such a high elasticity with good resistance to corrosion and high electrical conductivity as the beryllium bronzes.

On hardening from the indicated temperatures the structure of beryllium bronzes consists either of only grains of the solid alpha solution (e. g., the Bp. B2 grade bronze) or of a mixture of grains of the solid alpha solution with a small amount of grains of the solid beta solution (e. g., grade Bp. B2.5). In tempering (or artificial ageing) a highly dispersed chemical compound CuBe precipitates from grains of the alpha solution. The beta solution decomposes in tempering into a mixture of two phases, alpha solution and the same chemical compound CuBe.

It should be noted here that such operations as spring winding from a wire or blanking and bending of leaf springs from a ribbon must be performed before tempering, for the beryllium bronze in as-hardened condition is a very plastic material.

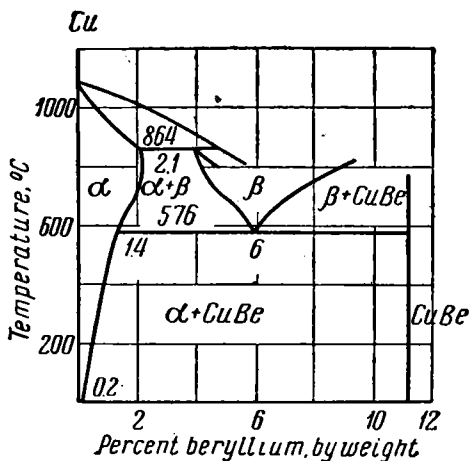


Fig. 168. Portion of phase diagram for copper-beryllium alloys.

*Chapter XIX***HEAT-TREATMENT OF ALUMINIUM-BASE AND
MAGNESIUM-BASE ALLOYS****92. ESSENTIALS OF HEAT-TREATING ALUMINIUM-BASE ALLOYS**

The heat-treatment of articles made of aluminium alloys is considered one of the most important operations in the whole manufacturing process. Its role is extraordinarily great in imparting high mechanical properties to these alloys. In this respect aluminium alloys are second only to steels. The aluminium alloys have become the

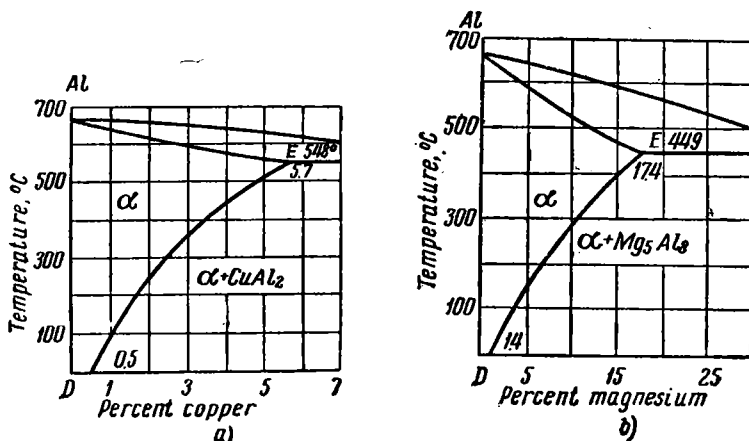


Fig. 169. Portions of aluminium-copper (a) and aluminium-magnesium (b) phase diagrams.

most important materials (after steels) used for structural purposes in the modern machine-building industry solely through heat-treatment. As structural materials the aluminium alloys even have some advantages over steels. In particular their specific strength, i.e., the tensile strength as referred to the specific gravity, is higher than that of steels, with the result that in some instances the weight of aluminium structures may be less than that of steel structures with the same strength and rigidity.

The heat-treatment of aluminium alloys is based on the fact that the solubility of many elements in solid aluminium decreases with rising temperature. Fig. 169 gives typical phase diagrams for binary alloys of aluminium with copper and magnesium. From these phase diagrams it will be seen that the solubility of copper in its solid

solution with aluminium decreases from 5.7 per cent at a temperature of 548° C to 0.5 per cent at room temperature, and that of magnesium from 17.4 per cent at 449° C to 1.4 per cent at room temperature.

Alloys of aluminium with a copper content up to 0.5 per cent are evidently not capable of being hardened, nor are aluminium-magnesium alloys with the magnesium content lower than 1.4 per cent. On the contrary, aluminium-copper and aluminium-magnesium alloys containing copper or magnesium in excess of 0.5 or 0.4 per cent respectively are capable of materially changing their phase composition and, consequently, structure on heating and cooling.

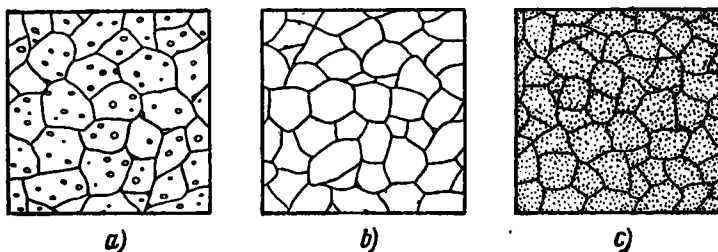


Fig. 170. Sketch illustrating the change in structure of an aluminium alloy hardened by heat-treatment.

The transformations in structure of an aluminium alloy with a constitution diagram similar to those represented in Fig. 169 is schematically shown in Fig. 170. In as-annealed condition this alloy (Fig. 170, *a*) has a structure consisting of solid alpha solution (solid solutions of copper in aluminium, or magnesium in aluminium, etc.) and small precipitates (usually of a spherical shape) of the second phase (chemical compositions of aluminium with copper, CuAl_2 , or of aluminium with magnesium Mg_2Al_3 , etc.). In heating to a temperature above the *DE* line indicative of the maximum, or limit solubility (Fig. 169), the alloy becomes single-phase, for the precipitates of the second phase will be fully dissolved in the alpha solid solution with the result that the structure of the alloy will consist only of grains of this solid solution (Fig. 170, *b*). If this alloy is then rapidly cooled (or quenched), its structure will remain what it was at a temperature above the *DE* line. Heating the alloy to some temperature below the *DE* line of limit solubility (see Fig. 169) will give rise to a repeated precipitation of the second phase (Fig. 170, *c*). The phase composition of such an alloy in as-tempered condition is the same as it was in as-annealed condition; however, the microstructures of annealed and tempered alloys are by no means

similar, differing from each other in degrees of dispersion, since the precipitates of the second phase in as-tempered condition are many times finer than those in as-annealed condition.

This is why a hardened alloy is strengthened by tempering, i. e., as the fine precipitates of the second phase form in great quantities within grains of the solid solution they progressively inhibit the intracrystalline slips, possibly by keying the slip planes, or by setting up strains in the lattice. Quenched aluminium alloys harden especially markedly in tempering, when the second phase is a chemical compound (as in the case of aluminium-copper or aluminium-magnesium alloys).

From a consideration of the phase and structural transformations which take place in heating and cooling the aluminium alloys it is also necessary to draw the following conclusions. It is possible, even without making experiments, to assume that the ductility of aluminium alloys is at its highest level in as-quenched condition, when their structure consists of grains of the solid solution only. This assumption is fully corroborated by conventional shop practice; e. g., to carry out operations of cold-working, such as bending, drawing, and riveting, blanks of these alloys have first to be hardened.

It is very important to pay attention to the fact that aluminium alloys do not recrystallise completely on heating. This implies in turn that if the alloy has been overheated and its structure has become coarse-grained, it cannot be corrected, i. e., restored or made fine-grained again by any heat-treating operation available. Thus, overheating cannot be corrected in aluminium alloys at all. In this respect the heat-treating operators dealing with aluminium alloys are certainly in more difficult position than those who work with steels. From this it is necessary to draw the most important conclusion, based upon practical experience, i. e., in the heat-treatment of aluminium alloys temperature control must be provided with especially great precision.

93. HEAT-TREATMENT OF WROUGHT ALUMINIUM ALLOYS

Of the great number of wrought aluminium alloys available (there are about 20 standardised grades alone), alloys of the duralumin type are the most important.

Duralumin is supplied in several grades, the principal ones being Д1, Д16, and Д3П. In these grade designations the letter "Д" indicates duralumin and the letter "П" that it is plastic, while the figures designate the number of the alloy, having no other physical sense at all. Duralumins are alloyed with three elements: copper, magnesium and manganese,

Grade of duralumin	Per cent copper (by weight)	Per cent magnesium (by weight)	Per cent manganese (by weight)	Average total content of alloying elements, per cent
Д1	3.8 to 4.8	0.4 to 0.8	0.4 to 0.8	5.5
Д16	3.8 to 4.9	1.2 to 1.8	0.3 to 0.9	6.5
Д3П	2.6 to 3.5	0.3 to 0.7	0.3 to 0.7	4.1

These three grades of duralumin differ from each other in content of alloying elements, the least alloyed being grade Д3П, and the most alloyed grade Д16. The more highly alloyed the metal, the more capable it is of increasing in strength. On the other hand, the

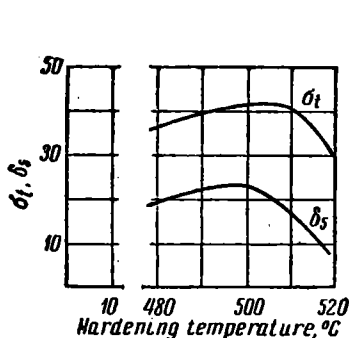


Fig. 171. Diagram showing the effect of hardening temperature upon the mechanical properties of grade Д1 duralumin which has been hardened and artificially aged.

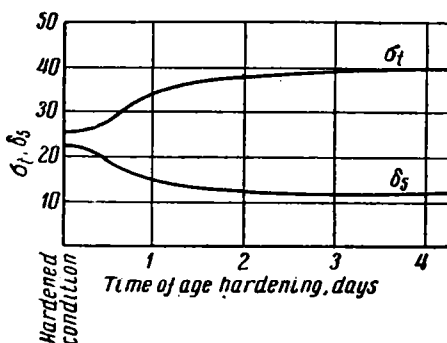


Fig. 172. Diagram showing the change in mechanical properties of hardened duralumin which has been naturally aged.

lower the content of alloying elements, the more plastic is the duralumin. Duralumin of grades Д1 and Д16 is used to manufacture rods, sheets and tubing. Wire for rivets is made from the Д3П grade duralumin.

The duralumin-type alloys are subjected only to one heat-treating operation, quenching. The heating of duralumin for quenching is conducted either in batch-type furnaces or in saltpeter baths. The quenching temperatures are accepted as follows:

Grade of duralumin	Hardening temperature, °C
Д1	505 to 510
Д16	495 to 505
Д3П	490 to 500

The diagram in Fig. 171 shows the effect of the hardening temperature upon the mechanical properties of grade Д1 duralumin. This diagram shows that duralumin-type alloys are very sensitive to overheating, and at the same time emphasises the necessity of close

control of the temperature of hardening. The holding time of heating at the hardening temperature is selected on the basis of the size of cross-section of the article to be treated. Blanks and articles are quenched in water at 30 to 40° C, the temperature of the quenching water increasing with the degree of intricacy of the shape of the article to be treated.

One of the most outstanding properties of the duralumin-type alloys is that on storing at ordinary temperatures after the hardening treatment described above they undergo a spontaneous change,

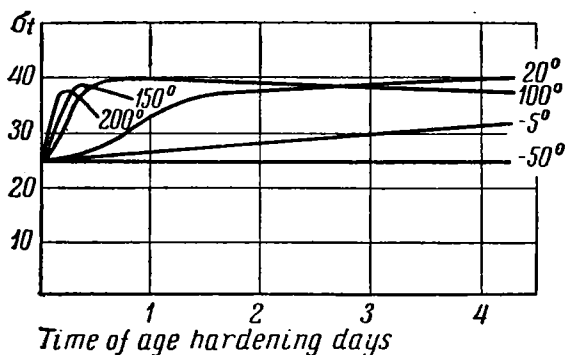


Fig. 173. Diagram showing the effect of the age-hardening temperature upon the change in tensile strength of hardened duralumin.

commonly referred to as ageing, which involves further increases in strength, hardness and yield point, with little change in elongation but with a decrease in plasticity (Fig. 172). The change is at first rapid but proceeds at a gradually decreasing rate. It is largely effected in the first few hours and is generally considered to be substantially complete in about four days. The process so outlined is known as the natural age hardening of duralumin-type alloys.

At low (sub-zero) temperatures the change in mechanical properties of duralumin proceeds at a considerably slower rate (Fig. 173). On the other hand, at elevated temperatures (as in artificial ageing) the alloy changes in its properties at markedly increased rate compared with natural age hardening; though heating the alloy to so high temperatures increases the rate of ageing, it has but little effect on the mechanical properties attained after several hours which begin to decrease after the maximum value of the tensile strength has been reached.

The mechanism of the natural age hardening of the duralumin-type alloys was unexplained for a long time. The pattern of strength-

ening that has been considered above (see Fig. 170), which is applicable for all aluminium alloys, has proved to be unsatisfactory for explaining the strengthening which takes place in the natural age hardening of duralumin. Neither microscopic nor X-ray examinations have succeeded in detecting the presence of precipitate of the second phase in a hardened and naturally aged duralumin. The structure of duralumin which has been hardened and naturally aged apparently remained single-phase. What then is the actual cause of the strengthening process occurring in duralumin?

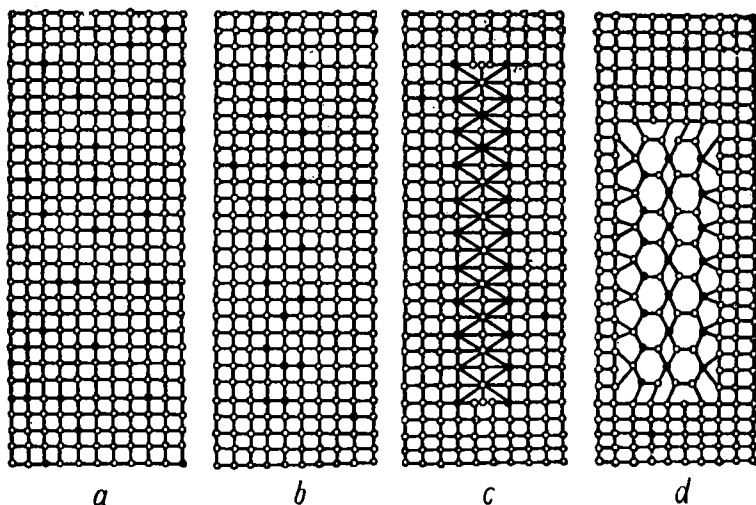


Fig. 174. Diagram showing the changes in the space lattice of hardened duralumin during the age-hardening process.

a—hardened condition; b—formation of Guinnet-Preston zones (local enriching of space lattice in copper atoms); c—inside the lattice of aluminium begins the formation of a lattice of chemical compound CuAl_2 ; d—end of the formation of the lattice of the chemical compound at the instant prior to its separation from the lattice of the solid solution.

In recent times it has been found that in the natural age hardening of duralumin-type alloys certain *intraphase* transformations take place. The nature of these transformations is illustrated by a very comprehensive diagram, prepared by A. P. Gulyaev (Fig. 174). In as-quenched duralumin, i. e., in duralumin immediately after quenching, the space lattice of aluminium is characterised by atoms of copper, magnesium and manganese distributed statistically uniformly (in diagram Fig. 174, a, for the sake of simplification, are shown the atoms of aluminium and copper only). Gradually the copper atoms begin to concentrate in certain portions of the space lattice (Fig. 174, b). In this process some portions of the solid solution

grains are depleted of copper, whereas other portions become markedly enriched in it. These portions of the space lattice of the solid solution are called Guignet-Preston zones, after the names of the scientists who first discovered this phenomenon.

As a result of such a concentration of copper atoms in separate portions of the space lattice it undergoes a marked distortion, its parameter in the Guignet-Preston zones being considerably greater than in other parts of the space lattice. The distortions of the space lattice

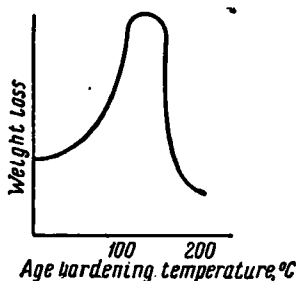


Fig. 175. Diagram showing the effect of the temperature of artificial ageing upon the corrosion resistance of duralumin.

set up great internal stresses, leading to the progressive hardening of the metal.

The intraphase transformations consist simply of the formation of the Guignet-Preston zones in the course of the natural age hardening of duralumin; no other transformations can occur, no matter how long the duralumin is kept at room temperature. Heating the alloy to a temperature of 150 to 200° C results in further development of the intraphase transformations. The second stage of the age hardening process then occurs, involving the regrouping of atoms in the Guignet-Preston zones, which process corresponds to the formation of

the chemical composition CuAl_2 (Fig. 174, c). When these portions of chemical compound have grown to a definite size (usually up to 3,000 Å long and 100 Å thick), they detach from the space lattice of the solid solution, forming new crystalline formations; this is the second phase in the duralumin structure (Fig. 174, d). This formation of independent crystalline precipitates in the second phase denotes the third stage of the age hardening process.

As soon as the independent crystalline precipitate of the second phase has been formed, it begins to coagulate. This is the fourth and final stage of the artificial age hardening. As is evident from the diagram in Fig. 173, the coagulation of grains of the second phase results in a gradual decrease in the strength, elasticity and hardness of the alloy, its structure gradually approaching that of an annealed one.

Thus, the artificial age hardening of duralumin detrimentally affects its mechanical strength. The corrosion resistance of duralumin-type alloys is also decreased by the formation of the double-phase structure which takes place in the artificial ageing process (Fig. 175), with the result that duralumin-type alloys which are naturally aged possess higher corrosion resistance than those artificially aged. From the theoretical point of view this is quite logical, since

under otherwise equal conditions a single-phase alloy with a homogeneous structure is always more resistant to corrosion than an alloy with a non-uniform or heterogeneous structure.

To increase their corrosion resistance, duralumin-type alloys in sheet form are *clad* in the course of hot rolling with a thin coating of high-purity aluminium on each side. These products, known as alclad alloys, possess an excellent resistance to corrosion.

According to the U.S.S.R. State Standard on the alclad alloys, the thickness of each layer of high-purity aluminium should be not less than 4 cent of the total thickness of the duralumin sheet when it is not heavier than 2.5 mm, and not less than 2 per cent of the total when the duralumin sheet is over 2.5 mm thick. In the heat-treatment of alclad duralumin products the holding time of heating at the hardening temperature must be as short as possible, because a long holding is likely to result in a diffusion of copper from the duralumin into the high-purity aluminium coating, thus decreasing the corrosion resistance.

Besides duralumin-type alloys, the wrought aluminium alloys capable of being hardened by heat-treatment include also the following three groups:

- 1) Avial of grade AB (aluminium-silicon-magnesium-copper-manganese alloys);
- 2) complexly alloyed forging alloys of grades AK6 and AK8 (aluminium-copper-magnesium-silicon-manganese alloys), possessing good ductile properties at elevated temperatures;
- 3) complexly alloyed piston alloys of grades AK2 and AK4 (aluminium-copper-magnesium-silicon-nickel-iron alloys);
- 4) high-strength alloy B95 (aluminium-zinc-magnesium-copper-manganese-chromium alloys); the B95 grade alloy is not yet included in the U.S.S.R. State Standard for wrought aluminium alloys.

All these alloys are used in an artificially aged condition. The natural age hardening of these alloys either is not effective or takes a very long time, e. g., the natural age hardening of the B95 grade alloy is still continuing 60 days after the quenching.

In principle these alloys have just the same mechanism of age hardening as was described in the case of duralumin-type alloys.

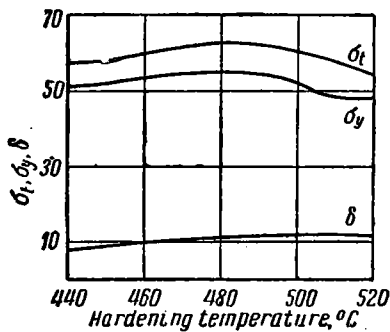


Fig. 176. Diagram showing the effect of the hardening temperature upon the mechanical properties of grade B95 alloy after quenching and age hardening.

They differ only in that the process of concentrating atoms of alloying elements in definite portions of the space lattice of the solid solution (the process of forming the Guignet-Preston zones), which takes place in the duralumin-type alloys at room temperatures, occurs and is completed in the case of alloys of grades AB, AK, and B95, mainly on heating to elevated temperatures, i.e., in artificial ageing. The second difference is that the artificial ageing of duralumin markedly accelerates the process of coagulation of the grains of the second phase, with the alloy growing mechanically weaker, whereas in the alloys of grades AB, AK and B95 the coagulation either does not take place at all or proceeds very slowly, with the result that these alloys are capable of being hardened by the artificial ageing process.

The heat-treatment of alloys of grades AB, AK, and B95 differs from that for duralumin-type alloys essentially only in the hardening temperatures and by the fact that the first are subjected to artificial ageing.

Grade of alloy	Hardening temperature, °C	Temperature of artificial ageing process, °C	Time of artificial ageing, hours
AB	515 to 525	150 to 160	6
AK2	510 to 520	165 to 175	15 to 18
AK4	510 to 520	165 to 175	15 to 18
AK6	505 to 515	150 to 160	12 to 15
AK8	460 to 475	135 to 140	16
B95 (not clad)	460 to 475	120 to 125	24
B95 (alclad)	500 to 510	175 to 180	8

The artificial ageing process may be conducted either in oil baths or, better still, in ordinary heating furnaces with an artificially circulated air atmosphere being forced to circulate.

Of all the aluminium alloys which are capable of being hardened by artificial ageing, grade B95 is the most important, because this alloy is the strongest known at the present time. The U.S.S.R. State Standard for rods of aluminium alloys specifies that in the quenched and artificially aged condition they must have the following minimum values of mechanical properties:

Rod diameter, mm	Tensile strength, kg/mm ²	Yield point, kg/mm ²	Elongation, per cent
Up to 22	50	38	7
23 to 160	54	41	6
Over 160	51	41	5

A considerable technical advantage of the B95 grade alloy is that unlike all other aluminium alloys it is only slightly sensitive to overheating. The diagram in Fig. 176, plotted according to data supplied

by I. N. Friedlander and E. I. Kutaitseva, shows the effect of the hardening temperature on the mechanical properties of this alloy. It is sufficient to compare this diagram with that of Fig. 171, showing the same relationship for the D1 grade duralumin, to become convinced of the superiority of the B95 grade alloy.

94. HEAT-TREATMENT OF ALUMINIUM ALLOY CASTINGS

A great number of cast aluminium alloys are available; the U.S.S.R. State Standard specifies some 22 grades. They are designated by the letters AJI (indicative of cast aluminium alloy), and by one or two-digit figures indicating the conventional number of the alloy, as:

Grade of aluminium alloy		Content of main alloying elements, per cent
AJ12	silumin-type alloys	Silicon 10 to 13
AJ14		Silicon 8 to 10.5; Magnesium 0.17 to 0.30; Manganese 0.25 to 0.50
AJ19		Silicon 6 to 8; Magnesium 0.2 to 0.4
AJ11		Silicon 6 to 8; Zinc 10 to 14; Magnesium 0.1 to 0.3
AJ12		Copper 9 to 11
AJ17		Copper 4 to 5
AJ16		Silicon 4.5 to 6; Copper 2 to 3
AJ15		Silicon 4.5 to 5.5; Copper 1 to 1.5; Magnesium 0.35 to 0.6
AJ13		Silicon 4 to 6; Copper 1.5 to 3.5; Magnesium 0.2 to 0.8; Manganese 0.2 to 0.8
AJ18		Magnesium 9.5 to 11.5
AJ13		Magnesium 4.5 to 5.5; Silicon 0.8 to 1.3; Manganese 0.1 to 0.4
AJ11		Copper 3.75 to 4.5; Magnesium 1.25 to 1.75; Nickel 1.75 to 2.25

The existence of such a great variety of cast aluminium alloys is explained, naturally, by the fact that not a single one of them possesses all the optimum properties. The best casting characteristics are peculiar to silumins AJ12, AJ19, AJ14 and AJ11. Aluminium-copper alloys of grades AJ12 and AJ17 have the highest machining characteristics. Aluminium-copper-silicon alloys of grades AJ16, AJ15, and AJ13 combine good casting characteristics, inherent to silumins, with good machining properties, characteristic of aluminium-copper alloys; however, their casting characteristics are inferior to those of silumins, while their machining properties are inferior to those of binary aluminium-copper alloys. Aluminium-magnesium alloys of grades AJ18 and AJ13 possess the best resistance to corrosion. The AJ11 grade alloy is the only heat-resisting cast aluminium alloy.

Almost all cast aluminium alloys are subjected to heat-treatment. Only two cast aluminium alloys, AJ12 (double silumin) and AJ11

(silumin, containing zinc), are not sensitive to this treatment. Castings of cast aluminium alloys may be subjected to the following types of heat-treatment:

Designation	Heat-treating operation
T1	Ageing (without preliminary hardening)
T2	Annealing
T4	Hardening
T5	Hardening and partial ageing
T6	Hardening and full ageing to obtain maximum hardness
T7	Hardening and stabilising tempering
T8	Hardening and tempering to soften the metal

Ageing without preliminary hardening (T1) is applied to castings of aluminium alloys of grades AJI3, AJI4, and AJI5, preferably produced by the permanent mould process. The permanent mould castings are hardened in the course of the rapid chilling of the metal in the mould.

Annealing (T2) is applied to some castings made of aluminium alloys of grades AJI3 and AJI6 which are required to have permanent shape and dimensions.

Hardening without subsequent artificial ageing (T4) is performed on castings of grades AJI7 and AJI9 aluminium alloys to increase their ductility (at the expense of a decrease of tensile strength and hardness), and on grade AJI8 aluminium alloy castings to make them still more resistant to corrosion. These cast aluminium alloys are not liable to natural ageing.

Most castings of aluminium alloys AJI1, AJI3, AJI5, AJI7, and AJI9 are subjected to *hardening followed by a partial ageing*. Such a heat-treating process produces practically the highest values of tensile strength, yield point and hardness that can be developed in cast aluminium alloys. This type of heat-treatment is the most commonly carried out on castings of aluminium alloys.

Hardening followed by full ageing to develop maximum hardness (T6) is used less frequently. This type of heat-treatment is applicable to large castings of the AJI4 grade aluminium alloy.

Hardening with subsequent stabilising tempering (T7) is applied to those castings of grades AJI3 and AJI5 aluminium-copper alloys which are intended to work at elevated temperatures.

Finally, *hardening and tempering to soften the metal* (T8) is carried out to improve the ductility of small castings of the AJI3 grade cast aluminium alloy.

Table 29 summarises the processes of heat-treatment of castings of cast aluminium alloys and the minimum allowable values of mechanical properties developed in them by these treatments (specified by the U.S.S.R. State Standard).

Table 29

Processes of Heat-Treatment of Castings of Aluminium Alloys

Designation or symbol of heat-treating process	Grade of alloy	Hardening		Age-hardening or annealing		Minimum mechanical properties		
		Temperature, °C	Holding time, hours	Temperature, °C	Holding time, hours	Tensile strength, kg/mm ²	Elongation, %	Brinell hardness
T1	AJ13	—	—	175 to 185	5	17	1	70
	AJ13	—	—	175 to 180	15	20	1.5	70
	AJ15	—	—	175 to 185	15	16	—	65
T2	AJ13	—	—	280 to 300	2 to 4	12	—	65
	AJ16	—	—	280 to 300	3	15	1	45
T4	AJ17	510 to 520	10 to 15	—	—	20 (21)	6	60
	AJ18	430 to 450	15 to 20	—	—	28	9	60
	AJ19	530 to 540	12	—	—	18 (19)	4	50
T5	AJ11	510 to 520	2 to 4	210 to 230	2 to 4	20	0.5	95
	AJ13	520 to 530	4 to 6	175 to 185	5	21 (24)	0.5	75
	AJ15	520 to 530	4	175 to 185	5	20	—	70
	AJ17	510 to 520	10 to 15	145 to 155	2 to 4	22 (23)	3	70
	AJ19	530 to 540	12	145 to 155	1 to 3	20 (21)	2	60
	AJ14	525 to 540	2 to 6	170 to 180	15	23 (24)	3	70
T7	AJ112	510 to 520	10 to 15	145 to 155	2 to 4	17	—	100
	AJ13	520 to 530	4 to 6	225 to 235	5	20	1	70
T8	AJ15	515 to 530	4	225 to 235	5	18	1	65
	AJ13	495 to 505	5 to 6	325 to 335	3	18	2	65

Note: The values of tensile strength indicated in brackets are given for permanent mould castings, and those above the brackets, for sand castings. Only one figure is given when the specified values of tensile strength are the same both for permanent mould castings and sand castings.

In analysing the figures of Table 29 attention is drawn to the long holding times of heating of some cast aluminium alloys for quenching. This may be explained both by the low rate of dissolving of the intermetallic compounds in the solid solution and by the necessity for homogenising, which is much slower and longer for cast aluminium alloys than for wrought aluminium alloys, where it is generally at least partly performed in the course of the hot mechanical working of the metal.

In hardening the metal is most frequently quenched in water at 50 to 100° C. The quenching water must be heated to avoid the development of quenching cracks in the castings of cast aluminium alloys, the majority of which usually have fairly low ductility (see Table 29). The temperature of the quenching water should be selected nearer to the upper limit indicated as the intricacy of shape of the castings to be quenched increases.

After artificial ageing or annealing the aluminium castings are cooled in the open air.

95. HEAT-TREATMENT OF MAGNESIUM ALLOYS

Six wrought magnesium alloys are available (grades MA1, MA2, MA3, MA4, MA5, and MA8), and six magnesium cast alloys (grades MJ11, MJ12, MJ13, MJ14, MJ15, and MJ16). The U.S.S.R. State Standard specifies the chemical composition and physical properties

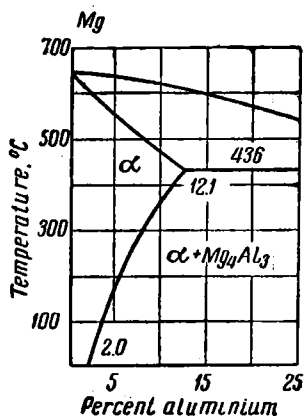


Fig. 177. Portion of magnesium-aluminium phase diagram.

of cast magnesium alloys. The best mechanical properties have been found in the MA5 grade wrought magnesium alloy and in the MJ15 grade cast magnesium alloy. For this reason these two magnesium alloys are most extensively used in industry.

Both these grades, MA5 and MJ15, are alloyed with the same alloying elements: aluminium (7.8 to 9.2 per cent in the wrought magnesium alloy, and 7.5 to 9.3 per cent in the cast magnesium alloy), zinc (0.2 to 0.8 per cent), and manganese (0.15 to 0.5 per cent).

A portion of the magnesium-aluminium phase diagram is represented in Fig. 177. A consideration of this diagram will show that this portion of it is similar to that represented in Fig. 169. Therefore the main principles of heat-treatment of magnesium-aluminium alloys are similar to those of the heat-treatment of aluminium alloys.

Magnesium alloys, both wrought and cast ones, are subjected to three types of heat-treatment: *annealing* (T2), *hardening* (T4), and *hardening with subsequent artificial ageing* (T6). Wrought magnesium alloys are annealed (T2) to recrystallise the metal and to increase its plasticity; cast magnesium alloys are annealed to relieve the internal stresses set up in solidification and cooling.

It is clear from Table 30 that the heat-treatment of magnesium alloys is characterised by a long holding time of heating at the temperatures of hardening and artificial ageing.

Magnesium alloys may be heated for quenching in vacuum furnaces, in batch-type furnaces with a protective gas atmosphere, and in baths with molten potassium and sodium bichromates. The protective gas atmosphere used is a mixture of air with 0.7 to 1.0 per cent (by volume) of sulphur dioxide. The magnesium alloys may be also heated in batch-type furnaces with an ordinary air atmosphere, but in such cases lumps of pyrite have to be placed in the furnace along with the articles to be heated.

Table 30 shows the values of mechanical properties of magnesium alloys in heat-treated condition.

Table 30

Processes of Heat-Treating Magnesium Alloys

Designation or symbol of heat-treatment	Grade of alloy	Hardening			Artificial ageing or annealing		Minimum allowable mechanical properties		
		Temperature, °C	Holding time, hours	Cooling	Temperature, °C	Holding time, hours	Tensile strength, kg/mm ²	Elongation, %	Brinell hardness
T2	MA5	—	—	—	350 to 380	3 to 6	25	9	50
	MJ15	—	—	—	170 to 250	3 to 5	15	2	50
T4	MA5	410 to 420	4 to 12	hot water	—	—	27	6	55
	MJ15	410 to 420	12 to 16	air	—	—	22	5	50
T6	MA5	410 to 420	4	hot water	170 to 180	16 to 24	30	5	60
	MJ15	410 to 420	12 to 16	air	170 to 180	16	23	2	65

